

# Transactions of *American Society for Steel Treating*

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*Interior of Blast  
Furnace Cast House  
in Central Alloy Plant*

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# AGATHON ALLOY STEELS



# TRANSACTIONS

## *American Society for Steel Treating*

VOL. XI

APRIL, 1927

NO. 4

### SPRING SECTIONAL MEETING

**P**RACTICALLY all arrangements have been made for the Spring Sectional Meeting to be held in Milwaukee, May 19 and 20, 1927, and the Milwaukee chapter committees are to be complimented for the earnest way in which they have carried out this work of preparation.

An excellent technical papers and plant inspection program has been prepared and this meeting gives every indication of equalling previous sectional meetings. The Hotel Pfister will be the headquarters.

Members will begin to gather on Wednesday, May 18, as on that day the officers and directors of the society will hold a meeting at the Hotel Pfister to transact important business of the society. The Recommended Practice Committee will also hold a meeting this same day. The customary Early Birds' Dinner will be held Wednesday evening.

Thursday and Friday will be devoted to technical sessions and plant inspection trips. Papers will be presented by Messrs. J. V. Green, R. S. Archer, H. B. Northrup, R. M. Sandberg, T. McLean Jasper and P. C. Osterman. The titles of these papers will be published next month.

Plant inspection trips have been planned which include many plants in and around Milwaukee. Among these plants are Allis-Chalmers Mfg. Co., International Harvester Co., and the A. O. Smith Corporation. Many other plants will be open for the inspection of our members. A complete schedule of events will be published in the May issue of TRANSACTIONS.

### METHODS OF DETERMINING HARDNESS COMPARED

**I**T is with much interest that we have received Technologic Paper 334 of The Bureau of Standards in which the relationship is given between the results obtained by two widely used methods of measuring the hardness of metals. In this investigation the Brinell and Rockwell methods were compared, and

tests were conducted on a great number of ferrous and nonferrous metals and alloys. Convenient semi-experimental formulas have been derived, by means of which the Rockwell or the Brinell hardness number can be computed from the other number with an error of less than 10 per cent. Similar formulas were obtained also for tensile strength, which can be computed from the Rockwell number within an error of 15 per cent.

The importance of this work will be realized when it is remembered that the safety of buildings, bridges, and machines of every kind depends upon the correct use of structural materials. The designer strives to be economical in the use of materials so as to avoid excessive weight and needless expense, yet each part must be sufficiently strong for any load which it may have to bear.

In order to find the strength of any metal or alloy a portion of it is worked into a convenient shape and tested in a testing machine. In most cases the designer is interested in the tensile strength of the material. A tensile test is, however, expensive and, often, no sufficient number of specimens to represent adequately the whole part, can be obtained.

Fortunately, a substitute for the tensile test, which is satisfactory for a great many practical purposes, exists in the so-called indentation hardness test. It is inexpensive, can be made in a few minutes, does not require an expensive preparation of the specimen and can often be made on a finished part without destroying it.

This test consists in making an indentation in the part which is tested, by a hard indenting tool loaded with a constant load. It is evident that the harder the material the smaller will be the indentation. Therefore, the size of the indentation is the measure of the indentation hardness. The indentation hardness number which expresses a certain relation between the constant load and the variable size of indentation may be used to calculate with a sufficient degree of accuracy the tensile strength of material.

Two of the most often used indentation tests in this country are the Brinell and the Rockwell tests. In order to be able to compare the hardness or the tensile strengths of two materials for one of which the Brinell and for the other the Rockwell number are known, it is necessary to know the relation between these numbers. This relationship has been worked out in the present investigation.

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## STRESSES IN QUENCHED AND TEMPERED STEEL

BY SAMUEL L. HOYT

### *Abstract*

*When a high-carbon steel is quenched from above its critical point in water, a transformation takes place in the steel at about 660 degrees Fahr. (350 degrees Cent.) which is accompanied by a volume change of the order of 1 per cent. It has been shown by Heyn, Scott, and others that the mechanism of the reaction of the steel to the quenching operation is such that internal stresses are necessarily set up. Tempering is accompanied by volume changes and we may assume that the internal stresses are materially altered before being relieved. Further, if the dimensions of the steel are followed throughout the quenching and tempering operations, it will be observed that they are not related to the volume changes by any simple law. It is the object of this paper to indicate how the stresses operate during quenching and tempering to produce strains in the steel and to show how, on such a basis, we may account for apparently anomalous dimensional changes. The general argument is that the tensile and compressive stresses may not, at times, exactly balance each other and that their tendency to equalize results in strains or dimensional changes which are superposed on those due to the volume changes.*

### MECHANISM OF STRESS FORMATION

THE mechanism whereby stresses arise during the quenching of steel has seldom received detailed treatment but Heyn, in his latest book, published after his death, has discussed this question in a manner which may well be considered here.(1)\*

If a steel is quenched from above its critical point, it takes on a greater volume than it would if allowed to cool slowly. This difference increases with the rapidity of the quench or, better, with the degree of martensitization. During the quench, the outer shell in contact with the cold water, cools more rapidly than the core so that if the union between outer and inner portions

\*The figures in parenthesis refer to the references appended to this paper.

A paper presented before the eighth annual convention of the Society, Chicago, September 20 to 24, 1926. The author, Dr. Samuel L. Hoyt, is a member of the research staff of General Electric Company, Schenectady, N. Y.



could be dissolved, the two would assume different volumes or different lengths. On account of the bond between the two, constraining them to assume the same length, the outer portion will be shorter than normal and the inner portion will be longer than normal. Expressed in terms of stress, the outer portion will be in compression and the inner portion will be in tension.

Passing to a mere detailed picture of the behavior during quenching, Heyn proceeds as follows: During the early part of the quench both parts of the bar are in the (temperature) zone of plastic deformation. The core cannot "give" to the compression of the outer shell, as the pressure is from all sides or simple hydrostatic pressure. This causes the outer portion to deform or lengthen and to accommodate itself to the inner portion. On entering the zone of elastic deformation (leaving the zone of plastic deformation) the deformation becomes elastic and hardening cracks may result. The core is still under compression on entering the zone of elastic deformation. It should now be recalled that the outer portion will end up with a greater specific volume than the inner, as it alone is converted into martensite. Between this point and room temperature a reversal in the volume changes must occur with the system passing through a stress-free state. Over this temperature interval the contraction of the core exceeds that of the outer shell and the stresses reverse in sign. That is, the outer shell comes under compression and the inner core under tension, when the bar reaches room temperature. The volume arrived at (final length) is a compromise and depends upon the thickness of the outer layer compared to the cross section, which in turn depends upon the diameter of the bar.

Following Heyn's reasoning, we may regard the difference in the shrinkage of the outer and inner zones as the cause of the signs of the stresses in those sections, and we should expect much larger stresses in plain carbon steels than in the deep-hardening steels of the low alloy type which become martensitic throughout.

Scott (independently) takes the argument farther and gives a much more thorough analysis of stress formation (2). Instead of discussing a general case as Heyn did, Scott differentiates between such factors as surface hardening and deep hardening, water quenching and oil quenching, large sized bars and small sized bars, etc.

He first notes that 1-inch round bars of a low-alloy gage steel

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are more apt to crack on oil quenching than on water quenching, but that water quenching smaller bars, which have the same temperature distribution as the 1-inch bar during oil quenching, likewise produced cracking. Bearing in mind that surface cracks are due to tensile stresses, his work showed that a reversal in sign occurred with increasing size or with the cooling power of the bath. The matter seemed to depend less on the magnitude of the stresses than on their sign and he attempted to account for his observations on the basis of the laws of stress generation during cooling. He then proceeds to consider the generation of stress in cooling glass, the expansion phenomena in steel during quenching, the temperature distribution during quenching, and finally, the quantitative estimation of the permanent stresses. Summarizing, he states that, "from previous studies of stress generation in cooling glass and steel it is shown that the permanent stresses described above are produced by plastic deformation during quenching," and further, that, "permanent tensional stress at the surface can be generated by cooling only when the steel is plastically deformed while the coefficient of expansion is negative." From this, it is quite clear that the dimensional changes, even though they are governed by the volume changes, need not accord with them. The difference will be due to the strain, either elastic or plastic.

It is largely due to the work of Scott that we have a reasonably complete understanding of stress formation in quenched steel. Others have suggested mechanisms of this action and have suggested methods for applying the results of their analyses to the prevention of quenching cracks, but our present purposes do not require further consideration of this subject here. (3, 4, 5.)

#### MEASUREMENT OF INTERNAL STRESSES

It is a matter of common knowledge among those who handle heat treated steel that quenching "sets up stresses" or "strains." These stresses or strains are said to cause warping or cracking and one of the objects of tempering is to relieve these stresses. The first increase in strength on tempering is supposed to be due to the release of the stresses. In spite of this general knowledge and the importance of the subject, we are almost without specific information in this field, with the exception of the work of Scott. One of the first questions which requires a quantitative an-

swer is in regard to the magnitude of the stresses. Are they large enough, for example, to cause distortion or cracking or even to produce slow flow? To get an answer to this question the writer, in the summer of 1925, measured the stresses in a 1-inch bar of Sanderson's high carbon tool steel which had been quenched in water from above its critical point. The method of Heyn was used to measure the stresses. This steel is a surface hardening steel and hardened to a depth of about  $\frac{1}{4}$  inch. The stresses found were high, of the order of 150,000 pounds per square inch at the maximum and varied from compression at the surface to tension at the center. The stresses measured were the axial stresses and no account was taken of the radial and tangential stresses. These results agree with those subsequently published by Scott (2) and need not be reproduced here. Scott has shown further that a similar bar of deep hardening steel, quenched under similar conditions, has the same stress variation but that the stresses are considerably smaller, or of the order of 50,000 pounds per square inch, at the maximum. This may be accounted for by the fact that the core of the deep hardening steel expands by martensitization, and thus releases the stresses somewhat, while the core of the surface hardening steel does not.

#### METHOD AND TECHNIQUE

For the sake of completeness, it may be well to add a few words about the technique employed to measure the stresses. The bar was quenched from about 1470 degrees Fahr. (800 degrees Cent.) according to the instructions given in the American Society for Steel Treating Data Sheet N-3. Immediately after quenching, the hardness was about 64 on Rockwell C scale, but this increased to about 66 in 1 hour and 20 minutes standing at room temperature. The stress variation during this change would be interesting to follow but no method is known to the writer by which stress measurements can be made in such a short time. The length and diameter of the bar were then measured in the constant-temperature room of the small tool and gage department at the Schenectady Works of the General Electric Co. according to the standard methods of the art. A layer of metal  $\frac{1}{32}$ -inch thick was then ground off the bar over almost the entire length. This was done on the precision grinder of the research laboratory, slowly and under a stream of water to avoid heating effects. The



bar was then returned to the constant-temperature room for another length measurement, and the cycle continued until the bar was reduced to a sufficiently small size to get the stress distribution over the cross section. Inasmuch as the bar was allowed to remain in the constant-temperature room over night to allow its temperature to come to normal, the measurements were extended over a period of two weeks time.

The stresses in each layer of metal were calculated then on the basis of a modulus of elasticity of 30,000,000 pounds per square inch. It may be of some interest to note that a 1-foot bar shortened 1/21-inch on being ground from 1 inch in diameter to 3/8-inch, due to the release of the stresses in the outer layers. This gives an idea of the constraint imposed on a bar of steel by quenching. The formulæ used in calculating the stresses from the data obtained may be secured from the original source. (6)

#### THE EFFECT OF STRESSES ON THE TRANSFORMATIONS

Observations have been made on quenched steel, some obvious and others not so obvious, which seem to indicate that stresses play an important (though subordinate) rôle in governing the transformations in steel during quenching. Thus Hanemann and Schulz made various microscopical observations on quenched carbon steel which seemed to require, for their explanation, the assumption that pressure had exerted a positive influence in determining the micro-constituents which formed and their relative amounts. (7) This assumption was also in line with the well-known volume changes which accompany the transformations. On tempering the quenched steels, an expansion was observed at about 390 degrees Fahr. (200 degrees Cent.) and this was accounted for on the basis of a release of stress. Volume changes also accompany the changes which take place on tempering and they concluded therefrom that stresses must likewise have an effect on tempering, or during tempering. One observation which stands out in this connection was the occurrence of martensite on the outside, troostite (called osmondite) on the adjacent inner shell, and troostite plus martensite on the inside of the troostite shell. There was no doubt that the rate of cooling decreased from the outside toward the center and the occurrence of the martensite on the inside was said to be due to a pressure effect which accompanied the transformations. Thus a careful microscopic ex-

amination of the microstructure of quenched steel required the utilization of Le Chatelier's principle to account for the observations.

Le Chatelier's principle tells us that stresses tending to decrease the volume (simple hydrostatic pressure) will oppose a constitutional change which is accompanied by an increase in volume, and further, that stresses tending to increase the volume (negative hydrostatic pressure) will assist such a transformation. In the austenite = martensite transformation, we have a reaction which proceeds with an increase in volume, and in the light of the experience of Hanemann and Schulz, we may expect that hydrostatic pressure will retard, and negative hydrostatic pressure will assist it, as the case may be. Lest there be a misunderstanding here, it may be well to point out that it is not suggested that pressure acts as a variable factor, such as temperature and concentration, in the physico-chemical sense. The pressures generated put the steel as a whole in an unstable condition and the reaction of the steel is simply its attempt to restore equilibrium.

It might be objected that the Le Chatelier principle is not applicable to the present case due to the fact that the equilibrium is not stable. See, for example, Benedicks, *Zeit. Phys. Chem.*, 1922, vol. 100, p. 42, who states that it holds only for stable systems. The present writer, while agreeing in general with Benedicks, would hold that the Le Chatelier principle applies to the austenite-martensite system during elastic strain, or in other words, to part of the transformation in the outer shell and to all of the transformation in the inner core.

From the work of Scott on stress generation in deep hardening steel, we may divide the quenching operation into three consecutive steps. The first is the period between the commencement of cooling and the moment the outside layer has reached the temperature at which it begins to transform. The second period is the further cooling of the outer layer to room temperature, and the third is the cooling of the interior to room temperature. The sign and magnitude of the stress will then depend on the lag between the second and third periods. If the lag is small, the stress will be negligible. As the lag increases, the stress increases in magnitude to a maximum and then decreases and finally reverses in sign. For example, for a given  $A_r''$  transformation, as the temperature drop from center to outside increases, the stress at

the surface increases to a maximum, then decreases and finally changes to compression.

The application of the principle of Le Chatelier requires a knowledge of the stress condition during the transformation and the effect of the volume-change during the transformation on this stress condition. At the commencement, the outside is always in tension and the inside in compression. This follows from the argument stated earlier in this paper. However, these stresses will be limited by the elastic limit of austenite at this temperature and they may be neglected (Scott). Hence we may proceed as follows:

Except for very small temperature drops from center to outside, the outside layer goes under compression during  $A_r''$  due both to its own tendency to expand and to the shrinkage of the core. If the temperature drop is large, the tension on the inside will be quite large and the core will have the tendency to pass through  $A_r''$  under tension. This stress or negative hydrostatic pressure will assist rather than retard the transformation to martensite. If the temperature drop is medium, the (temporary) stress is correspondingly lower. At the start of  $A_r''$  the conditions will be about as in the former case with the outside going under compression and the inside under tension. On account of the small difference in temperature this condition is reversed and the inside goes under compression and the outside under tension. This compression is direct hydrostatic pressure, and we have a condition which opposes the transformation of austenite to martensite. If it now be recalled that the large difference in temperature between outside and inside corresponds to a water quench and the medium drop to an oil quench, we have a logical explanation of the larger amounts of austenite which are observed in oil-quenched steels. Scott (8) adds the requirement here that the martensite temper fully during the quench for the production of the maximum amount of austenite, and that it remain undecomposed for the minimum amount. Both of these views indicate the prominent effect of stresses (pressure) on the transformation of steel during quenching.

Another case of a pressure effect, though perhaps not quite so obvious, is that of a "deep hardening" steel. Such a steel becomes martensitic over the entire cross section when moderately small sizes are quenched. On the basis of cooling velocities and compositions alone, it would probably be impossible to account



for a martensitic center. Bearing in mind that negative hydrostatic pressure would favor martensite formation from austenite, we may logically conclude that its assistance is sufficient to produce the deep hardening observed.

From the point of view just advanced, it will be seen that at least two factors determine the hardness of quenched steel and the depth to which the hardening action penetrates—cooling rate and internal stresses. In a straight carbon steel the velocity of the austenite transformation is so great that the cooling rate controls. This applies particularly to the upper or  $Ar'$  transformation, during which austenite changes to troostite. Water quenching produces martensite only on the outside (in a 1-inch round bar) and troostite forms on the inside at  $Ar'$ . At the lower temperature, or at  $Ar''$ , the inside does not expand by passing through  $Ar''$  and no hardening can take place, even though strong tensile stresses are acting on the core. The addition of a small amount of retarding elements, sufficient to inhibit the troostite formation at  $Ar'$ , gives hardening on both oil and water quenching. Water hardening produces martensite on the outside while at the center the tensile stresses assist sufficiently to give deep hardening. With oil hardening the tensile stresses in the outside assist sufficiently to produce martensite while the compressive stresses at the center tend to retain austenite and to retard the transformation. The (slow) production of martensite will likewise give a deep hardening effect and the only difference between water hardening and oil hardening on the actual hardness number will be due to the larger amount of austenite and the more complete tempering of the martensite in the oil-quenched steel. This austenite would certainly have a tendency to change into martensite even at room temperature, and the increase in tension on the outside produced thereby may be held responsible for cracks which form after the steel has reached room temperature. Scott assumes that these cracks do not form as a result of increasing stress, probably because his stress analysis assumes that the stress reaches its maximum as soon as the temperature at the inside and the outside equalize.

It may also be noted that this point of view would require an oil-quenched steel to have a greater tendency to expand on standing at room temperature than a water-quenched steel, a point which is well confirmed by the work of Scott. It would also re-

quire the oil-quenched steel to show a greater increase in hardness than the water-quenched steel. This requirement has likewise been confirmed by the author and by the published data. The volume-change which accompanies these room temperature changes probably does not follow any simple law and will be considered later.

#### UNBALANCED STRESSES IN QUENCHED STEEL

We have already seen that the volume-changes and the accompanying stresses do not occur simultaneously in all parts of the steel. This fact gives rise to what we may term here "unbalanced stresses." The general subject of unbalanced stresses need not be gone into in detail but to get an understanding of what is meant by unbalanced stresses, we may consider the following. In 1918 von Wartenberg advanced an explanation of the elastic after-effect in metals which was suggested by his observation that single crystal metals do not show the after-effect. (9) His hypothesis is based on Tammann's assumption that flow or plastic deformation starts in polycrystalline metals in favorably oriented grains at a stress which is well below the elastic limit of the metal as ordinarily measured. At higher stresses some of the grains are strained elastically and some plastically, until finally the elastic limit is reached. On releasing the load, the elastically strained grains force the plastically strained grains to revert practically to their normal length during the elastic after-effect. The net result is that some of the grains are left in tension and others in compression. Inasmuch as this condition did not exist at the start of the after-effect, we may say that the stresses at the start of the after-effect were "unbalanced." This condition caused the plastic flow observed and such a flow or deformation is the natural consequence of the application of a stress which is not balanced. The writer realizes that this is an entirely inadequate explanation of the mechanism of the elastic after-effect and the rôle played therein by unbalanced stresses, but this will probably be sufficient to indicate the nature of the argument.

On measuring the dimensional changes which accompany the quenching and tempering of tool steel and noting similar data in the literature, the writer has been at a loss to account for the fact that the dimensional changes do not always follow the simple volume changes which are known to occur. Many steels extend in length and in diameter on quenching while others shrink

in length and expand in diameter. In either case the volume-change is that which would be predicted for the amount of martensite which forms. The only way to account for this anomalous dimensional change, in the writer's opinion, is to assume that plastic deformation has occurred. Plastic (or elastic) deformation implies that a force or stress is acting on the bar and inasmuch as such a stress would not produce deformation if it were balanced by a stress equal to it and of opposite sign, this implies unbalanced stresses during the deformation. The deformation is the reaction of the metal which tends to equalize the stresses.

The rôle which such unbalanced stresses play in the hardening of steel may be recognized by noting the manner in which the dimensions fail to follow the volume changes. One case has been given already, that of the dimensional changes on quenching. An example of this is to be found in steel No. B 46, Table II of Scott's paper (8). The length instead of increasing (with the volume) decreased 0.5 mils per inch while the diameter increased 1.3 mils per inch. A deep hardening steel, quenched in the same manner, increased both in length and in diameter. Tempering is also accompanied by volume changes and on the basis of a simple volume-change we would expect the length to change proportionately to the diameter. That it does not is clear from the data which indicate that a deformation, as if due to stress, has occurred. The dimensional changes recorded are, in the sense of this paper, the net results of the volume changes and of the unbalanced stresses which produce the distortion.

We have now to consider something of the nature of internal stresses and the reactions in quenched steel which may be considered to produce them. The elastic after-effect has already been mentioned and it has been pointed out that this effect is best accounted for as due to stress equalization which acts over a period of time. The reaction of the metal is a dimensional change. While it is probably true that similar redistributions of stress occur in quenched steel, optical methods would have to be used to detect the strains. The volume changes which accompany aging and tempering are more potent causes of distortion than that just mentioned. The elastic after-effect is touched on more to bring out the idea of unbalanced stresses than to indicate the cause of dimensional anomalies which are noted in quenched steel. This part of the subject will be returned to later, as it

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will be best here to consider the reactions which are held to be the principal sources of this effect.

#### TRANSFORMATIONS IN QUENCHED AND TEMPERED STEEL

It would be more satisfactory in the present discussion if certain definite and well-understood reactions could be used as the basis of the analysis of dimensional changes during quenching and tempering. That this basis is lacking is evident from Professor Sauveur's recent criticism of the theories of hardening (10) and the attempt here will be to draw as reasonable conclusions as can be from our available data.

The microconstituents of quenched steel are austenite, martensite and troostite. The austenite remains, as a general rule, as a result of an incomplete Ar" transformation and its amount depends upon the composition of the steel and the conditions of the quench. The transformation is not necessarily completed by the time room temperature is reached but may proceed afterward, as has been recently shown by Hanemann and Schrader (11). Heating a steel containing austenite to 212 degrees Fahr. (100 degrees Cent.) also allows Ar" to proceed further, provided only that it has not gone to equilibrium during the quench. If this remnant of Ar" be allowed to take place at constant temperature, we may expect it to be accompanied by an increase in volume and in hardness but inasmuch as this is only a minor or secondary transformation, we may assume that only subordinate effects will be accounted for by it. This partial reaction may be classed with the change from austenite to martensite at temperatures below room temperature as due to a metastable equilibrium between austenite and martensite, following Hanemann and Schrader, and may be considered as a quenching phenomenon.<sup>2</sup>

Coming to the tempering phenomena of austenite, we may assume that austenite changes rapidly into either martensite or troostite at around 390-570 degrees Fahr. (200-300 degrees Cent.). That austenite is responsible for this transformation was early shown by Maurer and is evident from the X-ray analysis of Heindelhofner and Wright (12). The X-ray evidence did not show conclusively whether martensite or troostite was formed. It has been assumed that martensite forms during this transformation

<sup>2</sup>Schottky, who observed heat being generated spontaneously in quenched steel when heated to 100 degrees Cent., ascribed this transformation to austenite = martensite. (*Ferrum*, 1913, vol. 10, p. 274.)

(Hanemann and Schulz, and Enlund) largely on the basis of the increase in volume observed, for it is well-known that martensite has a larger specific volume than austenite. Other writers concur in this opinion but apparently without attempting to consider the matter in detail.

Direct microscopic examination of a tempered austenitic steel shows that a troostitic structure forms during this reaction. This point has been shown by various writers, but is particularly well shown in a recent discussion by H. Schottky (13). This direct evidence is aligned with the volume-change by assuming that the martensite is so fine-grained that it looks like troostite. (Fraenkel and Heymann.) In view of this unsatisfactory situation it may be well to reconsider the evidence and to see if the direct observations are not confirmed by the physical data.

The assumption that the transformation is from austenite to martensite seems to hinge solely on the volume change, and this in turn on the assumption that an austenite = troostite transformation would not produce the expansion observed. This viewpoint seems to be incorrect as is shown, for example, by the greater density of austenite at  $Ar_1$  as compared to pearlite and its greater coefficient of expansion between  $Ar_1$  and room temperature.<sup>3</sup> This shows that it is not necessary to assume that martensite forms at the 570 degrees Fahr. (300 degrees Cent.) transformation on tempering. Furthermore, the transformation of austenite to martensite in sufficient quantity to account for the volume change, would necessarily raise the hardness, whereas the hardness actually decreases about 10 points on the Rockwell C scale. Finally, the expansion relates to steel which has been tempered for one hour (Scott) which makes it necessary to assume that martensite resists tempering to about that extent. This is not compatible with our common experience with quenched steel nor with the drop in hardness, and Fraenkel and Heymann have shown that the transformation is 70 per cent completed in 1 hour at 390 degrees Fahr. (200 degrees Cent.) and over 90 per cent completed at 680 degrees Fahr. (360 degrees Cent.). In this light it seems entirely logical to assume that the austenite transforms to troostite on tempering at around 390 to 570 degrees Fahr. (200 to 300 degrees Cent.).

<sup>3</sup>See, for example, Scott's steel S-10 quenched from 2012 degrees Fahr. (1100 degrees Cent.) (austenitic) which expands 0.002 inches per inch on tempering at 930 degrees Fahr. (500 degrees Cent.)

A similar "peak" is observed on tempering alloy steels as has been dealt with, for example, by Grossmann and Bain in their well-known work. Here the microscopic evidence is that austenite transforms into martensite, somewhat as it does in straight carbon steels at 100 degrees Cent. Confirming this direct evidence, we have an increase in both specific volume and hardness, which likewise indicates that martensite is formed. A difference between the two reactions just mentioned is that martensite transforms to troostite at 100 degrees Cent. in plain carbon steel while it does not do so in the alloy steels. The consequences of this state of affairs are quite definite but cannot be considered here.

It now remains to consider the entire course of the tempering curve. If we refer to the methods of the physical chemist in his study of reactions, and particularly to the use he makes of reaction velocities to reveal the mechanism of the reaction, we find that two procedures are followed. First, the change in the amounts of the reacting substances (reaction velocity) is followed as the reaction progresses with time. The temperature is kept constant during one such run. Other runs are made at different temperatures which cover the range which is to be studied. From such data, the reaction constant and the temperature coefficient of the reaction can be ascertained. To the writer's knowledge, Fraenkel and Heymann are the only experimenters who have made use of this procedure to study the transformations going on during tempering.

In metallographical investigations, it is customary to temper samples for some specific time interval (usually  $\frac{1}{2}$  or 1 hour) at different temperatures. The results of different tempering temperatures are then compared. The difficulty in interpreting such data is that the reaction has progressed to different stages at the different temperatures. For example, Fraenkel and Heymann show that the tempering of quenched steel has progressed 9 per cent at 78 degrees Cent.; 24 per cent at 99 degrees Cent.; 53 per cent at 160 degrees Cent.; 67 per cent at 200 degrees Cent. and 95 per cent at 360 degrees Cent. in 1 hour. The difficulties in interpreting such curves (Maurer; Hanemann and Schulz; Heindelhofner and Wright; and Scott, for example) are at once obvious, as are the possibilities of drawing varying and conflicting conclusions. This situation suggests the desirability of securing suit-



able data from which a quantitative analysis of the reactions can be carried out.

The general course of the tempering curve shows an increase in density and a decrease in the hardness and electrical resistivity. This change begins, undoubtedly, at room temperature, as is shown by the instability of quenched steel and the shrinkage experiment of Brush (14). A mathematical analysis of Brush's curve has failed to give positive evidence bearing on the nature of the reaction except that the property measured, shrinkage in length, does not reflect accurately the change in the amounts of the reacting constituents. With increasing temperature this reaction goes on with increasing velocity until, at about 355 degrees Fahr. (180 degrees Cent.), another reaction begins to make itself felt in the customary time of tempering. This reaction is accompanied by an expansion, according to measurements made at room temperature after tempering. (Enlund.) The question as to whether this reaction goes on at room temperature superimposed upon the former has not been satisfactorily answered, but Scott, after carefully considering the evidence, has concluded that it does not.

In view of what has just been recorded, it would be pertinent to inquire as to the reason for the increase in hardness observed on aging at room temperature (4). This is hardly to be considered as due to the same reaction that occurs at 180 degrees Cent. and may well be associated with a delayed  $Ar''$  reaction which is known to be quite sluggish at low temperatures. This is particularly true if the steel is removed from the quenching bath before it has reached the bath temperature. As Jeffries and Archer have pointed out, room temperature aging is accompanied by a decrease in volume and not an increase, which would have to be the case were the delayed  $Ar''$  the only reaction occurring. However, the martensite which has already formed must begin at once to change into troostite, for no incubation period has ever been observed, and we may ascribe the shrinking as being due to the transformation of martensite occurring simultaneously with the delayed  $Ar''$ .

Perhaps the best way to get a comprehensive view of the tempering curve is to consider it as a transformation of an unstable phase, such as "eta" of Hanemann and Schrader, into its constituents, and that the reaction varies in velocity with the temperature

at which it is allowed to proceed. At room temperature the velocity becomes so low, after a time, that even after many astronomical ages, it would be only half completed. At 285 degrees Fahr. (140 degrees Cent.) the reaction is half completed in 2 hours and at 390 degrees Fahr. (200 degrees Cent.) it is two-thirds completed in 1 hour. The chief constituent responsible for the reaction is known to be martensite but there is considerable doubt as to the constitution of the phases formed. Measurements of the electrical conductivity and density, which clearly indicate the endpoint of the reaction, point to a carbide,  $\text{FeC}_n$ , where  $n$  is about 10. On no other known basis can the variation in the reaction velocity with time be accounted for. Furthermore, tests for the carbide  $\text{Fe}_3\text{C}$  fail to indicate its presence until fairly high tempering temperatures are used.

Many writers have advanced views on the nature of martensite but few have dealt in any detail with the mechanism of the transformation of martensite on tempering. It is only with the latter that we shall deal at present. Thus Maurer, Rosenhain and Honda, among others, account for the properties of martensite, but their pictures would not enable one to predict the course of a tempering curve. The same may be said of the work of Hanemann and Schrader, although their work is indispensable to the proper understanding of some of the important phenomena. Jeffries and Archer, in their hypothesis, which accounts for the hardness of martensite as due to the small grain-size of alpha iron, are more explicit. According to their hypothesis, carbon atoms form cementite particles which thereupon grow by diffusion of carbon through the iron lattice. Diffusion of the carbon atoms is held to be the factor which controls the rate of the reaction, even at room temperature. As diffusion proceeds, the cementite particles increase in grain-size. Positive evidence either for or against this hypothesis would be difficult to adduce at this time due to the difficulties of dealing with grain sizes of the magnitude considered and the lack of specific information on the diffusion of carbon atoms through iron. However, such a hypothesis should be consistent with quantitative data and it may be well to consider certain facts in the light of a diffusion of carbon atoms.

The temperature coefficients of diffusion phenomena, based on a 10-degree Cent. rise in temperature, are known to be around 1 to 1.5. Fraenkel and Heymann found the temperature coeffi-

cient of the martensite transformation to be around 2.4, or so much higher than that of a diffusion phenomenon as to suggest to those experimenters that the rate of the reaction was not governed by diffusion. Not until a temperature of about 570 degrees Fahr. (300 degrees Cent.) is reached does diffusion affect the reaction velocity. This assumption is verified by the absence of an incubation period, for a transformation which depends upon nucleation for its start and diffusion for its completion, would start off differently than is indicated by the data. Furthermore, the heat of reaction, as calculated from the data of Fraenkel and Heymann, is 28,000 calories per gram mol. and for the temperatures involved is far too high to be consistent with a process of diffusion. Rather, it indicates a chemical reaction, the same as the high temperature coefficient does. The same objections would apply to Rosenhain's hypothesis for he assumes that the stable phases are already present in martensite. The changes which would occur on tempering would be physical, rather than chemical, and such an hypothesis is not consistent with the quantitative data.

The foregoing discussion places the transformations of quenched steel, whether the constituent be martensite or austenite, in the category of chemical reactions. On this basis they may be expected to obey the laws of physical chemistry. In particular the stresses which are present in the quenched steel may be expected to affect both the transformations proper and the rate of transformation in different sections.

#### UNEQUAL STRAINING DURING TEMPERING

A bar of quenched high carbon steel is not in a stable state, nor is it even in an unchanging metastable state if sufficiently accurate means are used to follow its changes. If the  $A_r$  reaction is not completed at the end of the quench some of the remanent austenite changes into martensite, at times over a considerable time interval. This has been shown by Matsushita for carbon steels (15) and is even more pronounced in alloy steels. In addition the martensite changes gradually into troostite although the velocity of the change soon becomes extremely small. The stresses in the steel are not uniformly distributed, and, what is more important here, they are not uniformly released after quenching. The microstructure is likewise not uniform throughout the quenched steel, so that on standing at room temperature or on tempering



at elevated temperatures different sections will react differently. Finally, we have an effect apparently due to stresses, which results in an appreciable transformation in the martensite before the austenite begins to transform. This is shown on the usual tempering curves by the expansion which comes in between two periods of shrinking, and on the constant temperature curves of Fraenkel and Heymann by a shrinkage which comes at the start, followed later by an expansion. Under these circumstances it is quite likely that volume changes will occur in different sections of the steel at different times and produce dimensional changes on tempering which do not accord with the overall volume changes. Viewed in this light some of the anomolous dimensional changes, or warping, can be readily accounted for.

A 1-inch bar of plain carbon steel quenched in water is martensitic on the outside and troostitic in the inside. The outside is under compression and the inside under tension. Tempering this steel tends to shrink the outside shell and therefore to release the compressive stresses, while it tends to release the stresses on the inside, but not to change its volume appreciably. Neglecting the stress effect, we see that tempering tends to decrease both the length and diameter of a round bar. Including the stresses, we see that the release of the compressive stresses in the shell which hold the tension on the inside in equilibrium, allows the bar to shorten and to expand transversely. This change in length is added to the shrinkage due to the volume-change while the transverse expansion is subtracted from the volume-change. From this we would expect the length to change more than the diameter and by an amount which is of the order of magnitude of the strain corresponding to the stresses involved.

Viewed in this light, the shrinkage of the shell dominates the dimensional changes, even during the expansion at the "peak." Referring to Fig. 1, (f) of Scott's paper (8) we see that the curve of the change in diameter lies above that of the length change. This fact is in close agreement with the view advanced here. Furthermore, the sharpness of the peak on the diameter curve is due partly to the martensite shrinkage and only partly to the austenite expansion.

The deep hardening, low alloy, steels are martensitic throughout when quenched in water in sizes of 1 inch in diameter. This steel behaves differently on quenching than a plain carbon steel

does, and increases in length. This failure to shrink would be associated with the low stresses set up and would be due to expansion of the core at  $A_r''$ . Generally, the length increases more than the diameter and this would be due to plastic extension of the outer shell during the early part of  $A_r''$ , although the initial condition of the steel and the manner of quenching also affect this, as has been shown by Storey (16). If the bar contained no austenite, tempering should cause a nearly uniform shrinkage, i. e., from martensite to troostite, except for the fact that the martensite on the outside (in compression) should tend to transform earlier. Tempering at 355 degrees Fahr. (180 degrees Cent.) for customary intervals (short, or for 1 hour) should cause nearly uniform shrinkage. Tempering at 510 degrees Fahr. (300 degrees Cent.) should show an expansion, but due to the relative concentration of austenite in the outer shell, the expansion should be greater there than on the inside and we may expect the length to change more than the diameter. How much the difference should be can hardly be stated, but the low stresses indicates that it would be quite small. We may also expect the "peak" to be more rounded, due to the earlier transformation of the core.

If the bar is quenched in oil, the stresses are just reversed, i. e., tension on the outside and compression on the inside (Scott). Here we would expect the martensite on the inside to transform first, being under compression, but we have just seen that the stresses are apparently too low to have an appreciable effect on the dimensional changes due to this source. If there were no austenite present, the shrinkage at 355 degrees Fahr. (180 degrees Cent.) should be uniform. It may now be recalled that this steel, when oil-quenched, contains an appreciable amount of austenite, sufficient to affect the volume-change or even to cause expansion at temperatures below 355 degrees Fahr. (180 degrees Cent.). From this we may reasonably expect a dimensional change from this austenite. During the quench, the outer shell passed through its  $A_r''$  when under compression, due to its own expansion and the shrinkage of the core. Afterward the core expanded and reversed the signs of the stresses, but we may conclude that during the  $A_r''$  reaction the outer shell has been largely under compression and the inner core largely under tension. This would produce more austenite in the outer shell than in the core, in spite of the stress condition in the fully cooled steel. Reasoning on this basis, we

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see that the tension in the outer core, by accelerating the austenite transformation, should expand the steel. This expansion would tend to lengthen the bar and to decrease its diameter. This strain would be superposed on the martensite contraction, and we should expect the length to decrease less than the diameter. At higher temperatures, the situation is more uncertain for there is considerable austenite present and it transforms over a wider range of temperature than usual.

This analysis of stress-effects during tempering, while incomplete, will serve to account for the conflict between volume changes and dimensional changes which are observed. One factor which cannot now be evaluated is the release of stresses due to temperature *per se*. However, the necessary consequences of this argumentation agree so well with Scott's measurements, that the theory advanced seems capable of answering the riddle of anomalous dimensional changes.

Manuscript received August 22, 1926.

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#### DISCUSSION

Written Discussion: By T. McLean Jasper, director of research, A. O. Smith Corp., Milwaukee.



The writer is not sufficiently familiar with internal stresses in high carbon steels to intelligently discuss the larger portion of this paper. He wishes however to suggest that if the cause of quenching cracks and seasoning cracks in metals could be fully explained this would have a far reaching effect in avoiding a considerable amount of grief which the manufacturer and user of metal is heir to.

The writer does not understand the reason for the author's statement that "the stresses may not, at times, exactly balance each other." This, it is assumed, refers to the sum of the forces locked up internally in a piece of metal when no external forces are applied. The velocity at which stresses act elastically in steel is so high that it would seem to the writer that this could not be substantially so. The fact that changes occur in the values of the internal stresses or in the volume of steel after quenching with slight changes of temperature or with time need not require such an explanation.

In a ductile material when the internal forces acting are sufficient to bring the metal at any point to the yield point stress there is a flow phenomena which occurs with time which can have the effect of relieving the stresses at this point very materially. This flow has been shown by Robertson\* to occur when once started at a much lower stress than the actual yield point. In this case cited the yield point stress was 14 tons per square inch or about 31400 pounds per square inch. Flow occurred at 11 tons per square inch to a strain value many times that necessary to bring the metal in the initial part of the test to the yield point. For this reason much relief of stress can be brought about and at the same time the balance of internal forces can prevail throughout the occurrence of relief.

The writer is of the opinion that the author has not sufficiently explained the phenomena which occur in steel when quenched and tempered. In using the method adopted by the author it is possible to find the stress distribution in hot rolling, cold rolling or in quenching and tempering steel by the assumption that with each layer of metal removed the summation of the remaining axial forces is zero and that the change of length of the remaining portion is elastically equivalent to the internal forces removed by the previous layer. It is only in simple cases that formulae developed by several investigators can be applied.

To measure the radial and tangential forces caused in a solid circular cylinder by quenching and tempering is a most difficult matter. This requires the most sensitive apparatus, the most ingenious methods and a complete application of the mathematical theory of elasticity.

#### Oral Discussion

J. M. LESSELLS: The paper we have just heard from Dr. Hoyt is certainly a very interesting one. It is very much to be regretted that such papers as these are not preprinted, because there is no opportunity to look at the subject in detail. This particular paper by Dr. Hoyt will give us no doubt a great deal of valuable information.

\*"On the Drop of Stress at the Yield Point of Ductile Materials." By A. Robertson. Complex Stress Committee Report. British Association, Toronto, 1924.

I would like to ask one question and then make a suggestion. The question is, if I understood the author correctly, he uses a test piece 12 inches long. I would like to ask what precaution he took to make sure there was no bending of the test bar itself. A test bar 12 inches long and say an inch in diameter is a very difficult thing to heat treat and keep the test bar perfectly straight. At the Westinghouse laboratories we have also been doing a little work on this question of residual stresses, and on Monday I presented a paper giving some data on the relation between Brinell hardness and tensile strength. I would like to suggest to Dr. Hoyt that in his further investigation, he include some data on the strength properties, not only looking at the problem from the metallurgical point of view, but looking at the problem from the mechanical point of view.

In this connection I would like also to draw his attention to the differences we observed on alloy and carbon steels. If we plot ultimate strength as ordinate and Brinell hardness as abscissa, as in the paper I presented on Monday I showed that under certain conditions we get a straight line relation between them. Taking the work of Gough of the National Physical Laboratories, and Professors Moore and Jasper for alloy steels, you will find that the point for the steel when quenched and tempered at the lower temperatures will fall above the line, but when tempered at the higher temperatures, it will fall on the line.

I would like to bring attention to the fact that this may be due to the effect of hydrostatic tensile stresses, on the Brinell. In other words, Dr. Hoyt has shown alloy steels to have a hydrostatic tensile stress. (I do not know whether he means this can be generally applied) but assuming that this is the case for the above particular steels, ( $3\frac{1}{2}$  per cent nickel and  $3\frac{1}{2}$  per cent nickel-chromium,) then the effect of the tensile stress on the Brinell comes up. In other words, if you have a hydrostatic tensile stress, the Brinell hardness may really be less, the material appearing softer than it otherwise would be, if there were no tensile stress present. With carbon steels, on the other hand, the points lie below the line. Here we have a hydrostatic compressive stress and would mean a higher hardness than is really the case.

O. Z. KLOPSCH: This paper of Dr. Hoyt's is very interesting. With carbon steel one inch in diameter, we know both from Scott's theoretical curves and from some actual curves that the cooling curve for the center would be something like this, the curve for the surface like this and other points in between. Therefore the outside fibers will pass through the  $A_r''$  range before the center gets to the  $A_r'$ . In other words, we will not have troostite at the center at the same time that martensite is formed at the outer fibers, for they will have already cooled through the  $A_r''$  range before the center has passed from the gamma condition. The condition that Dr. Hoyt discusses therefore is never really obtained at any one particular instant but rather over a period of time. I wished only to call attention to these facts.

A. V. DEFOREST: I would like to call attention to the hoop tension stresses in those quenched affairs, because in actual practice the tool always, or almost always, fails from hoop tension and not from the longitudinal stress.

ses, so that in the case where the outside is in tension, the cracks are longitudinal.

There are also a few observations possible from experience with shrinking dies that have a hole in the center. There the conditions are such that it is possible to suck the thing in by quenching it from the inside, or squeeze it in by quenching from the outside, and the relation of those two different methods of treatment is quite interesting in regard to the number of dies lost by cracks. I think that that method of experiment might also be taken into account rather than a solid cylinder.

A. L. DAVIS: Without trying to labor this point unduly, you may be interested in an example of distortion arising from repeated quenching, of a die of originally somewhat similar shape. It tended to approach the shape described by Dr. Hoyt, decreasing in one diameter and increasing in another in order to do so. In another case, that of a rectangular oblong piece, eventually the shape approximated the spherical, after several hundred requenches.

H. J. FRENCH: I think it is rather unfortunate that Dr. Hoyt's paper was not preprinted, and perhaps the question which I am going to ask him would not be necessary had we had an opportunity to see his results before this meeting.

I gained the impression from his presentation that he was able to quantitatively predict dimensional changes in steels from certain mathematical relations and some stress determinations and I would like to ask whether that impression is correct. My reason for asking that question is just this; in most computations of this sort there is involved at one stage or another consideration of the temperature drop. Investigators have so far largely made use of theoretical cooling curves in which, for water quenching, it is assumed that the surface instantaneously reaches the final cooling temperature. In checking some theoretical cooling curves over a very wide range of conditions, some years ago, with actual cooling curves obtained both in our own laboratory and in published reports, we found serious discrepancies. I wonder therefore whether these predictions by Dr. Hoyt are actually quantitative, and, if so, whether or not they represent special conditions. I am very much interested in finding out whether it is possible from such theoretical calculations to reliably predict dimensional changes.

There are also other factors which come into play, namely, the initial condition of the steel as affecting the final dimensional changes observed in hardening, the quenching temperature, and the condition of the water. I do not refer solely to the temperature of the water, but freedom from dissolved gas and other factors.

#### Author's Reply

To Professor Jasper, and Mr. French, I would say that I am simply trying to formulate a general picture of the effects of internal stresses on the peculiar dimensional changes, warping, etc., which are observed to occur in steel during quenching and tempering, and that the hypothesis is on a qualitative basis. We have only to assume that, for one reason or another, the volume changes occur in different parts of the steel at different times. The

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# OBSERVATIONS ON THE MICROSTRUCTURE OF THE PATH OF FATIGUE FAILURE IN A SPECIMEN OF ARMCO IRON

BY FRANCIS F. LUCAS

## Abstract

*This paper describes a co-operative study of the microstructure of the path of fatigue failure conducted under the auspices of the National Research Council.*

*The paths of fatigue failures in specimens of Armco iron are shown. Non-metallic inclusions act as "stepping stones" for the crack. Inclusions seem to influence the origin and trend of the crack. Grain boundaries do not appear to be a potential source of weakness.*

IN 1924 the National Research Council Division of Engineering requested the co-operation of Bell Telephone Laboratories (then operated as the Engineering Department of the Western Electric Co.) in the Fatigue Phenomena of Metals Investigation being conducted by Professor H. F. Moore at the Materials Testing Laboratory, University of Illinois, Urbana, Illinois.

The Metallographic Laboratory of the Bell System was requested to undertake a microscopic examination of the path of fatigue failure in certain specimens tested by Professor Moore with the object of determining the microstructure of the path of failure. The committee in charge wished to know how the fatigue cracks originate, how they progress across the specimen and the structural conditions which promote resistance to fatigue failure.

For the purpose of discussing the Laboratories' participation in the fatigue-of-metals activities of the National Research Council, a conference was held in New York early in 1925. There were present: Mr. W. Spraragen, Secretary—Division of Engineering and Industrial Research—National Research Council; A. D. Flinn, Director of the Engineering Foundation; Professor H. F. Moore; Dr. H. Clyde Snook, the Laboratories' representative on the

Acknowledgment is made to Bell Telephone Laboratories, Inc., for their courtesy in furnishing the half tone reproductions illustrating this paper.

A paper presented before the Chicago convention of the Society, September 20-24, 1926. The author, Francis F. Lucas, is associated with the Research Laboratories of Bell Telephone Laboratories, Incorporated, New York City.

National Research Council's Advisory Committee for the Fatigue of Metals Investigation; and other members of the technical staff.

The conference disclosed that little, apparently, was known concerning the actual mechanism of fatigue failures. Professor Moore had been studying various methods of approach which might disclose new information of value on the fatigue phenomena. He had designed several types of testing machines in which the specimen could be mounted in such a way that it could be examined microscopically without removing it from the machine. Unfortunately the limitations of high power metallography prevented the adoption of this method of investigation.

As a preliminary step it was decided that Bell Telephone Laboratories would use their facilities and personnel to examine a specimen of Armco iron which had been tested to destruction by Professor Moore. Armco iron was selected because it is the purest commercial form in which iron may be obtained. Such a specimen, it was thought, would offer the fewest number of structural complications and any information derived from a study of Armco iron would have some direct bearing later if the study of steels were undertaken. Guided by their observations on this preliminary specimen Bell Telephone Laboratories would make some further recommendations as to additional specimens. This plan was carried out and the results of the preliminary investigation were made available in report form to Professor Moore. At that time an additional specimen was requested which had been subjected to reverse cycles of stress sufficient to cause the appearance of a visible crack but not to cause final rupture. The reason for this will become apparent as the work of the Laboratories is discussed.

Both specimens have been carefully examined at high powers and the paths of failure in each case photographed at representative stages from the origin to the end of the crack. At the request of the committee the results of these studies are being made public for the benefit of those interested in the subject.

#### DESCRIPTION OF SPECIMENS

The general form and the approximate dimensions of the two specimens examined were the same. (Fig. 1).

The preliminary specimen was broken after the following cycles of stress: 1,313,600 cycles of reversed flexure with a maximum value of unit stress of 24,000 pounds per square inch; 1,875,200 cycles

of reversed flexure with a value of maximum unit stress of 25,200 pounds per square inch; and finally 115,300 cycles of reversed flexure with a value of maximum unit stress of 26,400 pounds per square inch.

The unbroken specimen with the approximate locations of the two cracks is shown in Fig. 1. The cracks were detected after 536,900 cycles of reversed flexure with a maximum value of unit stress of 25,800 pounds per square inch. The broken specimen described above failed in substantially the same place as indicated for the unbroken specimen.

The physical properties of the metal were given as follows:

Proportional Elastic Limit .....	16,100 pounds per square inch
Yield Point .....	19,000 pounds per square inch
Tensile Strength .....	42,400 pounds per square inch
Elongation in 2 inches .....	48.3 per cent
Reduction of Area .....	76 per cent
Brinell Hardness .....	69
Endurance Limit (rotating-beam test) .....	26,000 pounds per square inch

The chemical composition was reported as follows:

	Per Cent
Carbon .....	0.02
Silicon .....	0.02
Manganese .....	0.03
Phosphorus .....	0.005
Sulphur .....	0.042
Iron .....	Diff.

#### PREPARATION OF SPECIMENS

The preparation of the specimens for metallographic examination was in accordance with the practice described elsewhere and need not be repeated.<sup>1, 2</sup> Precautions were taken in each case to secure a flat surface out to the very edge of the specimen. The specimens were examined in planes parallel to the broad faces and each specimen was re-prepared a number of times. The structures were visually examined at varying depths from the surface inward. It may be said with reasonable certainty that the photomicrographs reproduced herewith are representative of the structures found throughout the depth and length of the cracks.

<sup>1</sup>High Power Metallography--TRANSACTIONS, American Society for Steel Treating, November, 1923.

<sup>2</sup>Appendix III, Sauveur, "The Metallography and Heat Treatment of Iron and Steel" (1926). Note by the writer of this paper on preparation of specimens.



### EXAMINATION OF PRELIMINARY SPECIMEN

The extreme edge of the path of failure was studied at high powers and from this examination the following conclusions were reached tentatively:

In relatively pure iron (Armco iron) the impurities must exert a profound influence on the behavior of the metal when subjected to reversed cycles of stress. Cracks resembling tears penetrate the metal from the path of failure. Grain boundaries apparently are not a source of weakness and, in fact seem not to have a controlling influence on the trend of the cracks. The failure follows a path marked by inclusions some of which have been dislodged.

From this preliminary study completed in April, 1925, it appeared desirable to have a specimen for examination which had not been tested to final rupture but only to a point where a well defined crack appears. Such a specimen can be prepared more advantageously for microscopic examination than a specimen with one broken face. In the former case the crack can be located midway of the specimen by suitably cutting the test piece, but in the latter case the broken edge of the specimen must be examined. This means that the specimen must be polished flat to the very edge of a rough surface.

Moreover, in the unbroken specimen the trend of the crack could be determined with certainty and without complications resulting from damage to the broken faces at the time of final rupture. Since the unbroken specimen confirms fully the conclusions deduced from a study of the path of failure in the broken specimen the latter will not be dealt with further at this time.

### EXAMINATION OF UNBROKEN SPECIMEN

Fig. 2 shows one of the broad faces of the specimen with the crack made visible by rather deep etching. As prepared for high power work the crack is so fine that it would be almost invisible at the magnification of 5.8 x used for this photograph. To overcome this difficulty, the cracks were developed by etching so that their location and penetration into the specimen would be evident. Only one crack is shown, the other is of similar configuration but extends inward from the edge diagonally opposite.

The photomicrographs reproduced herewith were taken of

the unbroken specimen and consist of two series. Since two cracks appeared in diagonally opposite edges, one series relates to a random section through one crack in a plane parallel to the broad faces of the specimen and the other to a similar observation plane through the diagonally opposite crack.

The plan followed was to photograph a series of representative fields from the origin to the end of each crack. The photographs are not a continuous series or panorama and discontinuities exist in some cases between succeeding photomicrographs. A few additional photographs have been taken to illustrate some details of structure such as the character of inclusions found in the metal.

The reproductions of the high power photomicrographs are arranged so that in each case the crack is proceeding from left to right. In some cases succeeding illustrations show fields closely adjacent; in others appreciable gaps intervene. The object has been to illustrate representative structures along the cracks.

For one series of pictures the specimen was etched lightly to avoid, as much as possible, widening of the crack. In these pictures grain boundaries have not developed. It was of importance to know whether grain boundaries influenced the trend of the crack and with the object of answering this question another series of pictures was taken with the specimen etched more deeply.

For the series of pictures, Figs. 3 to 17, the specimen was etched very lightly; therefore, the grain boundaries have not developed. The crack has been widened by the etching reagent but not to an appreciable extent. The specimen was etched with nitric acid. The magnification is 3500 x.

Fig. 3 shows the origin of the crack at the extreme edge of the specimen where apparently a large inclusion existed or there was an indentation of the metal. On some planes of observation the structure at the origin of the crack appears quite homogeneous: i. e., inclusions or indentations were not found in all planes. The crack progresses inward and passes around or through voids—in reality cavities formerly occupied by solid non-metallic inclusions. It was observed that in most cases the crack prefers as its path the boundary between the non-metallic inclusion and the metal, rather than to traverse the inclusion.

In Fig. 3, and in fact throughout the series of photomicrographs, many small black particles are seen. These are believed to

be inclusions of the same general nature as the larger ones which are more plainly seen and whose structure can be studied. In some cases small cavities are found and it may be that these cavities once were occupied by inclusions which during the process of preparation of the specimen became dislodged. That these small inclusions play a part in fatigue failures of the kind studied seems quite probable. It will be observed quite frequently that lateral cracks extending from the main crack end in these small inclusions or include many of them in their path.

In order that a true perspective may be had of the vision afforded by the high power pictures a few dimensions are given below. For instance in the series of pictures, Figs. 3 to 17, at 3500 x:

	Microns	Inch
The area included in the photomicrographs actually measures on the specimen about .....		0.0016x0.0011
The average diameter of the large inclusions (Fig. 16 for example) actually is about .....	5.	0.0002
The average diameter of the small inclusions visible as tiny black specks actually is about .....	0.14 (Min. about 0.07)	0.000,005
The width of the crack near its end (Fig. 13) is about .....	Max. 0.3 Min. 0.14	0.000,01 0.000,005

That the cavities result from inclusions dropping out during preparation was demonstrated in the course of the investigation. On one occasion, the specimen was re-prepared several times in succession and in one place (Fig. 27) the crack traversed the boundary between the metal and a large inclusion. Subsequently the inclusion dropped out during preparation and left a cavity such as these shown along the crack.

Whether the physical disturbance of the metal structure by the reversals of stress has loosened the bond between the metal and the inclusions in the path of failure or whether the inclusions are inherently seated loosely in the metal has not been fully determined. It does appear, however, that many of the cavities were observed to be along the path of failure if not actually in the cracked area.

Figs. 6 to 14 show gradual changes in structure of the crack from the origin to the end which is visible in Fig. 14. The struc-



tural characteristics have been discussed above so that it is not necessary to take up each picture in detail. The path of the crack is from one inclusion to another with here and there detours to one side or the other, which cannot always be accounted for by details of structure appearing in the plane of observation. It must be kept in mind that conditions either above or below the plane of observation may have exerted the controlling influence.

The end of the crack is shown in Figs. 13 and 14. Fig. 15 is of a field immediately adjoining so that the crack as it appears in Fig. 13 is trending directly toward the inclusion shown in Fig. 14. One of the inclusions shown in Fig. 15 is the same one shown in Fig. 14. If one examines the photographic print of Fig. 14, there will be noticed a suggestion perhaps of incipient cracks extending to the large inclusion.

A strained condition in metals, not shown by very light etching, is often revealed by deep etching, as is well known. This specimen was re-etched several times for short intervals of time and examined at high powers between successive etchings. Soon the crack developed and extended to the inclusion. The final condition is illustrated in Fig. 17 and to further clarify the illustrations the field shown by Fig. 17 has been indicated by dotted lines in Fig. 14. It seemed quite evident from this deep etching experiment that the metal all along the crack is in a strained condition. The width of this strained band of metal as measured by the eye in viewing the photographs appears quite appreciable but actually the zone shown in Fig. 17 including the crack is only two or three thousandths of an inch in width. It will be seen that the strained metal appears to flare out from each side of the crack in Fig. 17 and seems to be practically nil at the inclusion. Nearer the origin of the crack the width of the strained zone appears greater, but probably does not exceed a few hundredths of an inch. Observations seem to show that a fatigue failure is a highly localized condition.

Fig. 16 shows typical inclusions at random and without particular reference to the crack. The large white inclusions are believed to be iron carbide. It will be seen a little later that they stain with sodium picrate. These inclusions are not a source of weakness but seem to be a strengthening element since the crack avoids them throughout.

The dark inclusions are slate-blue in color but often they are not composed of a single constituent but are "tipped" with a light constituent or a black one. Sometimes all three constituents occur in the same inclusion. Occasionally the slate-blue constituent is "tipped" or partially surrounded with the lighter constituent and sometimes fully so. The light constituent is straw-yellow in color and is believed to be iron carbide possibly contaminated, as iron carbide usually is clear white. When these "tipped" inclusions occur along the crack, the crack seems to avoid the tipped end but does enter between the metal and the slate-blue constituent.

The light, straw-yellow constituent stains with sodium picrate in the same manner as iron carbide. Perhaps the best way to describe the manner in which the inclusions are held within the metal is to say that the metal seems to "wet" the carbide inclusions thoroughly but not the other inclusions.

In the series of pictures, Figs. 18 to 33 inclusive, the opposite face of the specimen was etched deeply enough to develop the grain boundaries. This, of course, widened the crack considerably and also developed incipient pitting of the iron. In this series the inclusions are much less abundant than in the series already described but in general the crack seeks the inclusions. The magnification for this series is 2170 x. The details of structure are much the same as those previously observed but it is quite apparent that grain boundaries are not a path of weakness since we find the crack crossing the grain boundaries, paralleling them, or following them with equal freedom. It may be said, I think, that if the grain boundaries are conveniently located and oriented the crack will follow them.

In Fig. 30 the end of the crack is shown trending toward a duplex inclusion. Just ahead, as shown in Fig. 31, a streak of inclusion is seen.

Figs. 32 and 33 are typical inclusions selected without reference to the crack. For these two pictures the specimen was etched with sodium picrate but for all others the specimen was etched with nitric acid. That the inclusions are composed of more than one constituent seems quite evident.

Visual examination of these specimens at high powers impresses one most decidedly that the ordinary chemical analysis does

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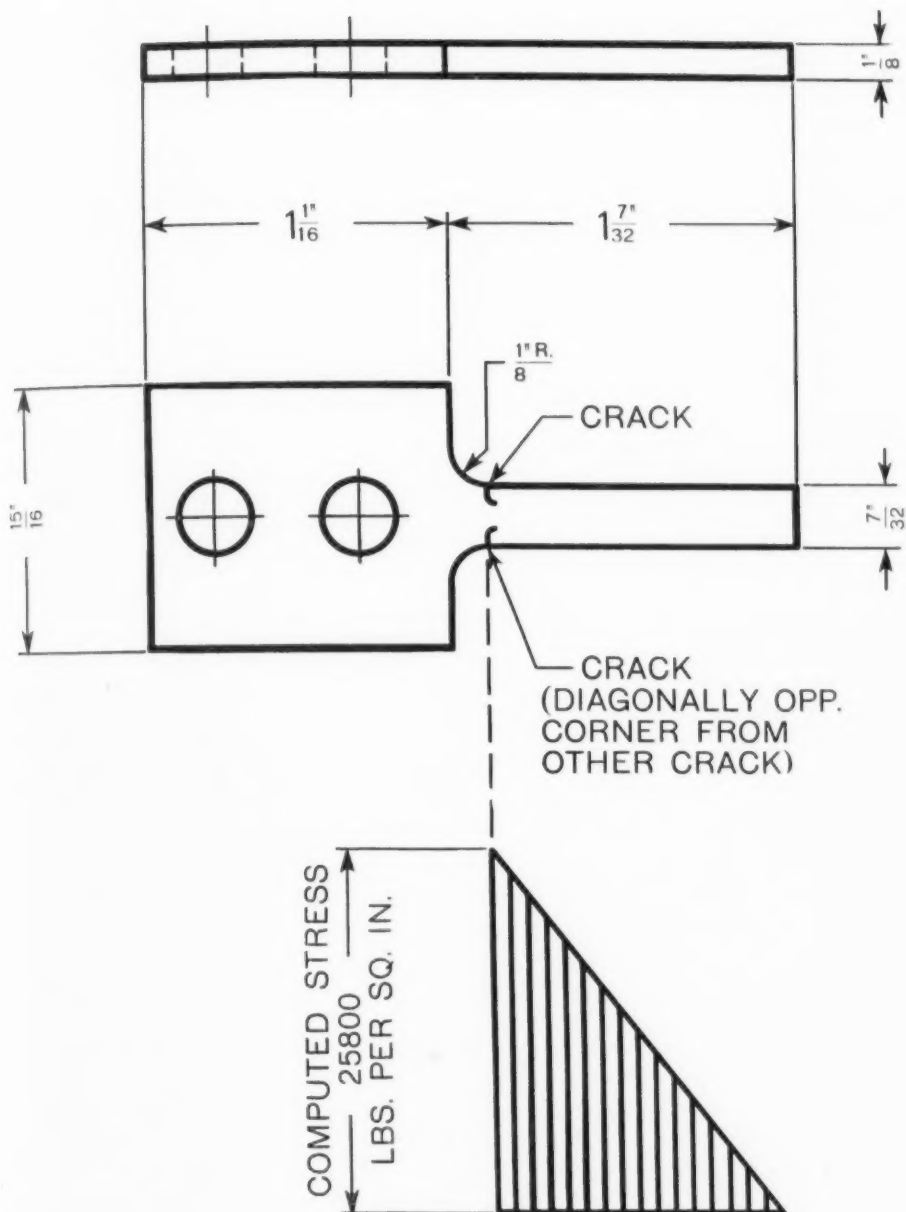


Fig. 1—General form and approximate dimensions of specimens examined. Computed stress and location of cracks refer to unbroken specimen







Fig. 2—A broad face of unbroken specimen, crack made visible by deep etching.  
Magnification 5.8X





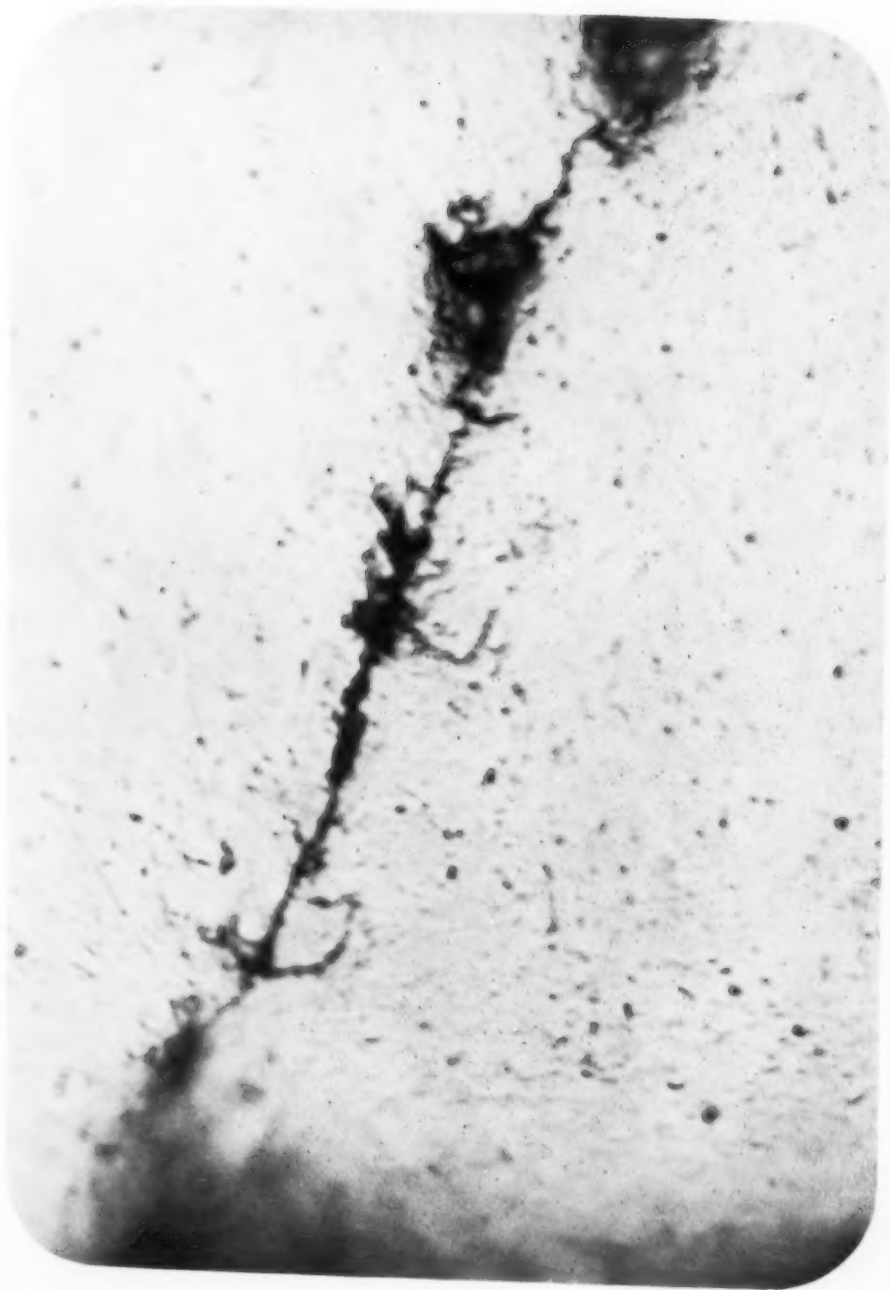


Fig. 3—Origin of crack. Magnification 3500X



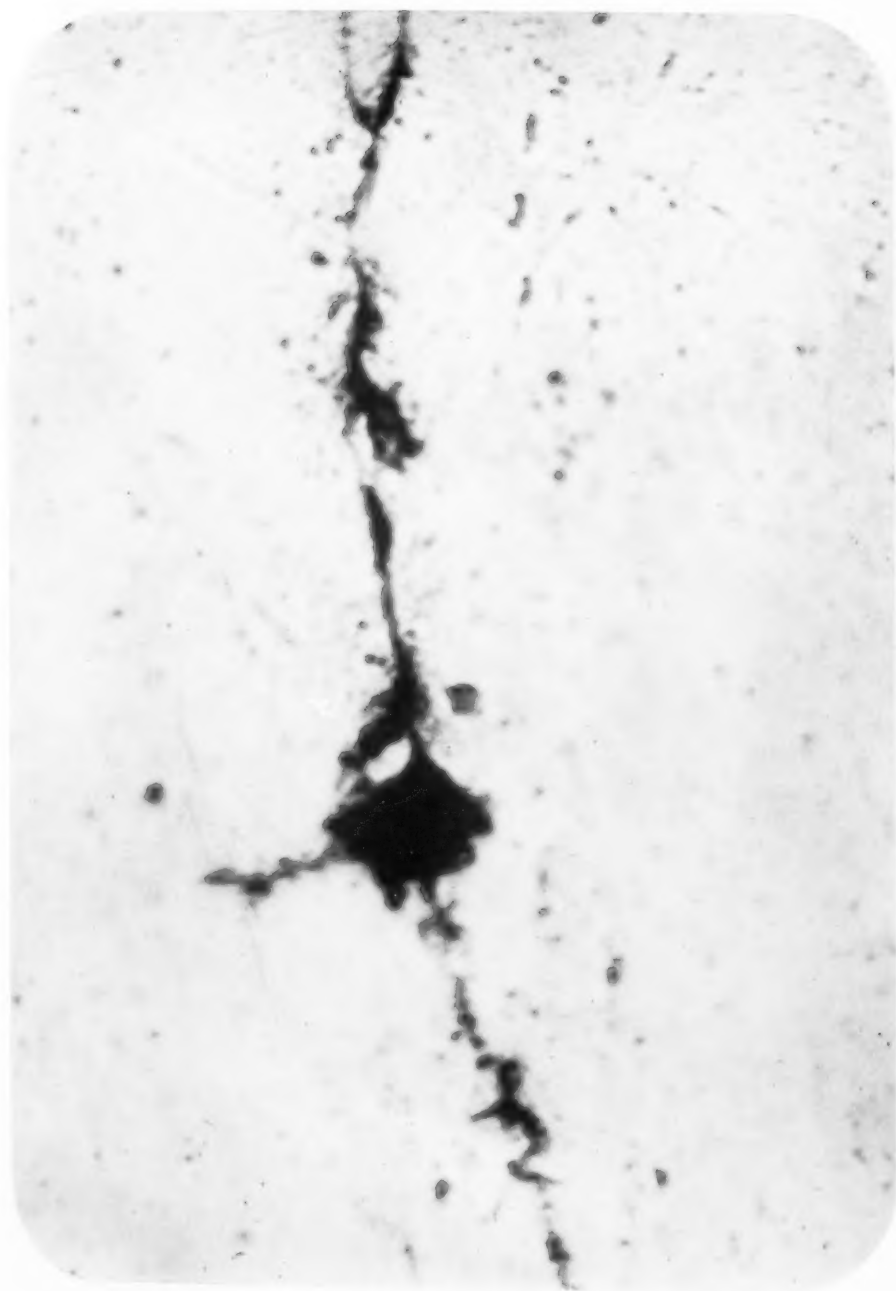


Fig. 4—Magnification 3500X





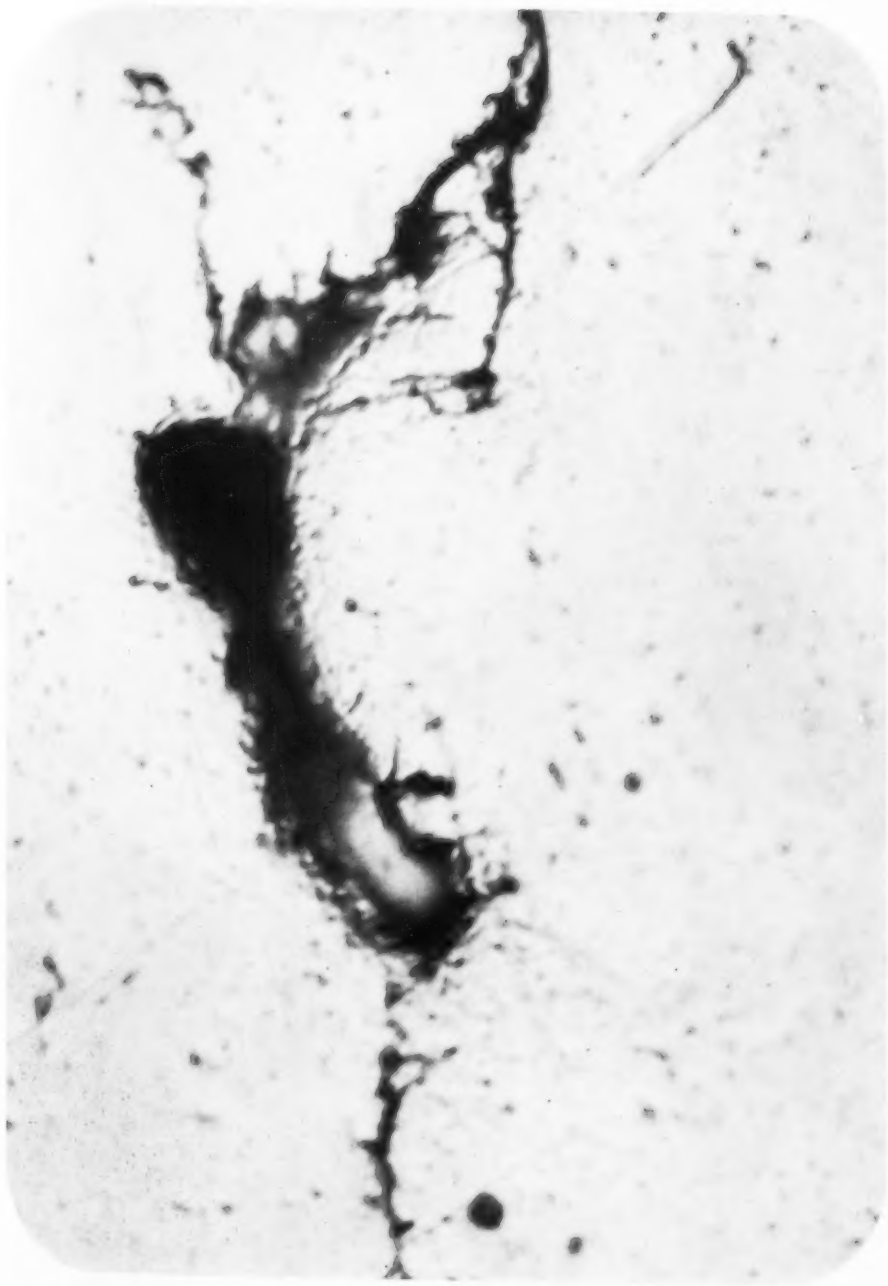


Fig. 5—Magnification 3500X



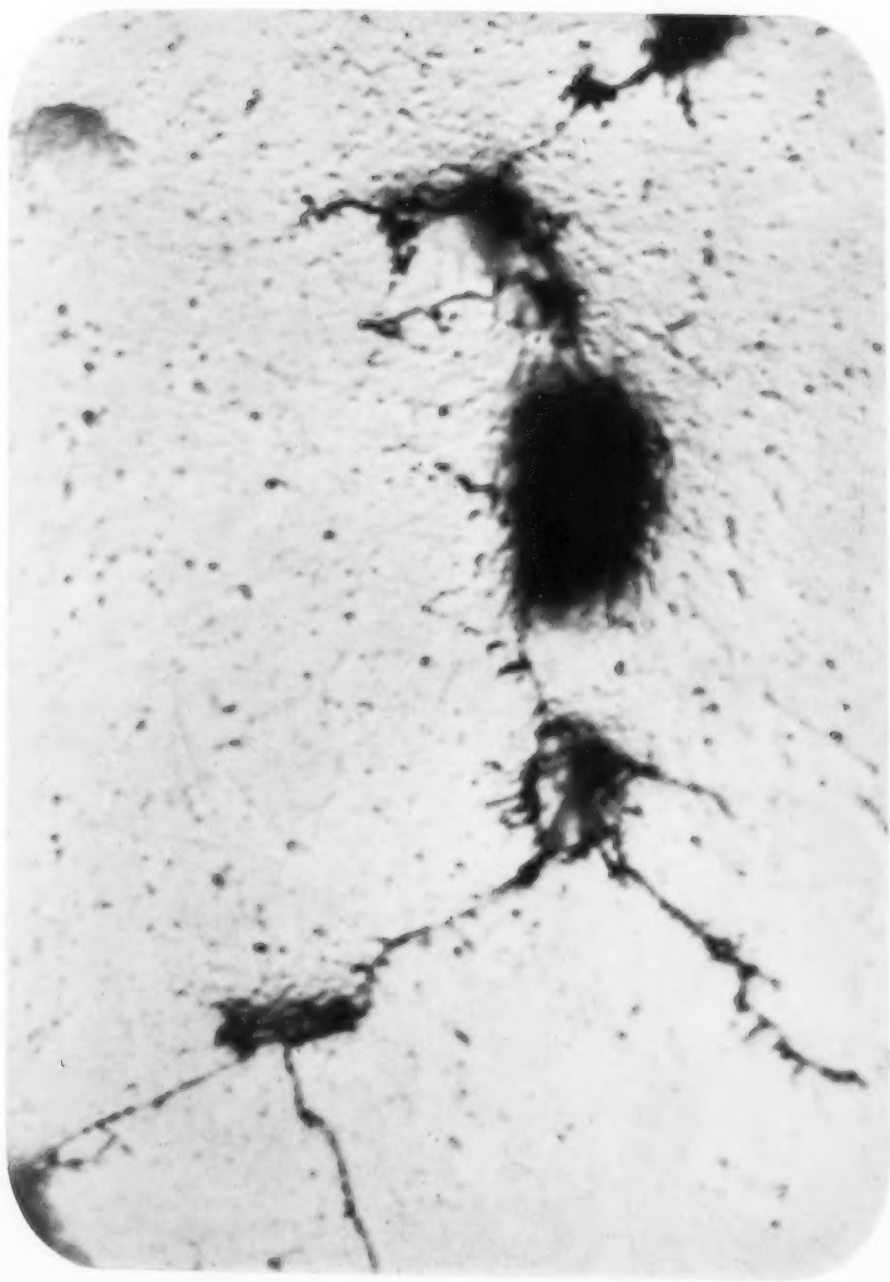


Fig. 6—Magnification 3500X





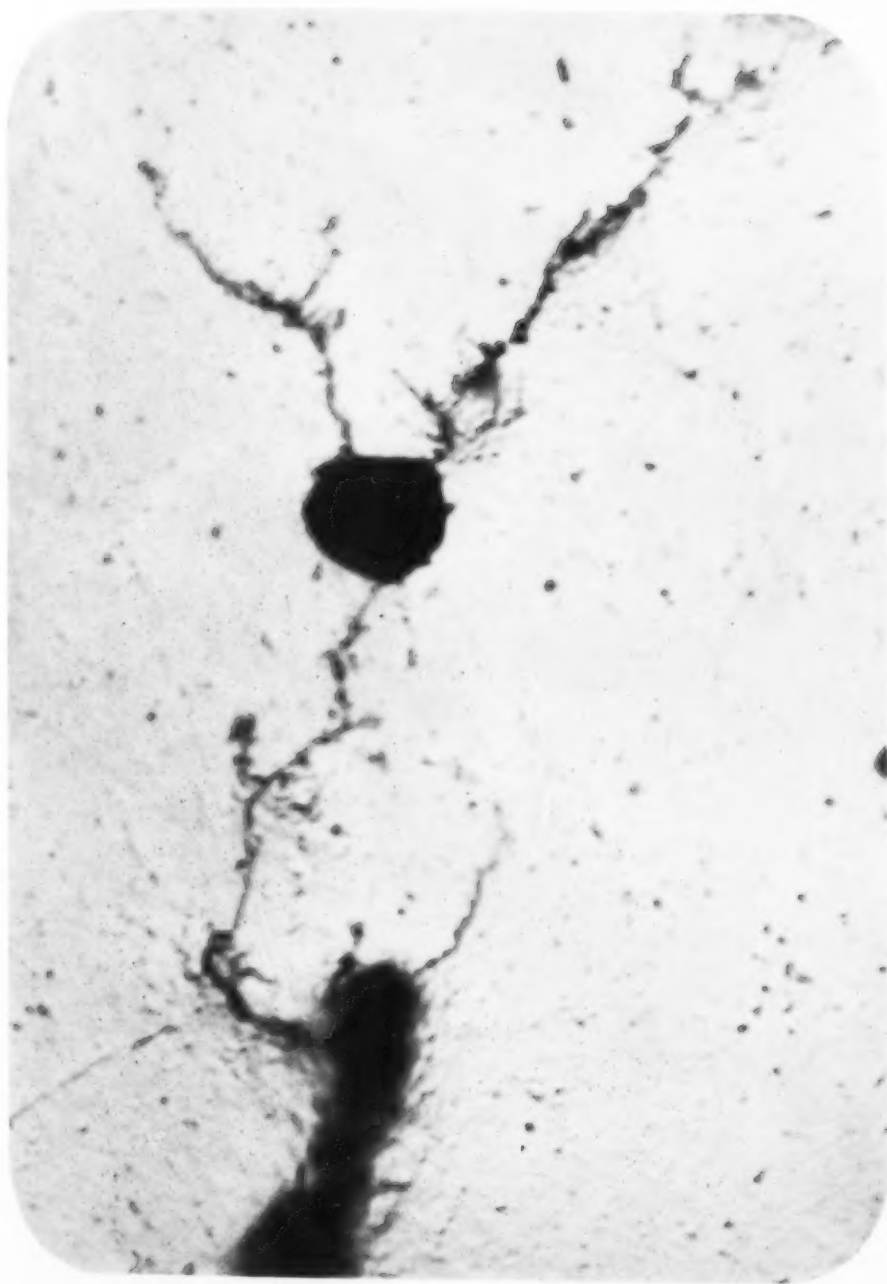


Fig. 7—Magnification 3500X



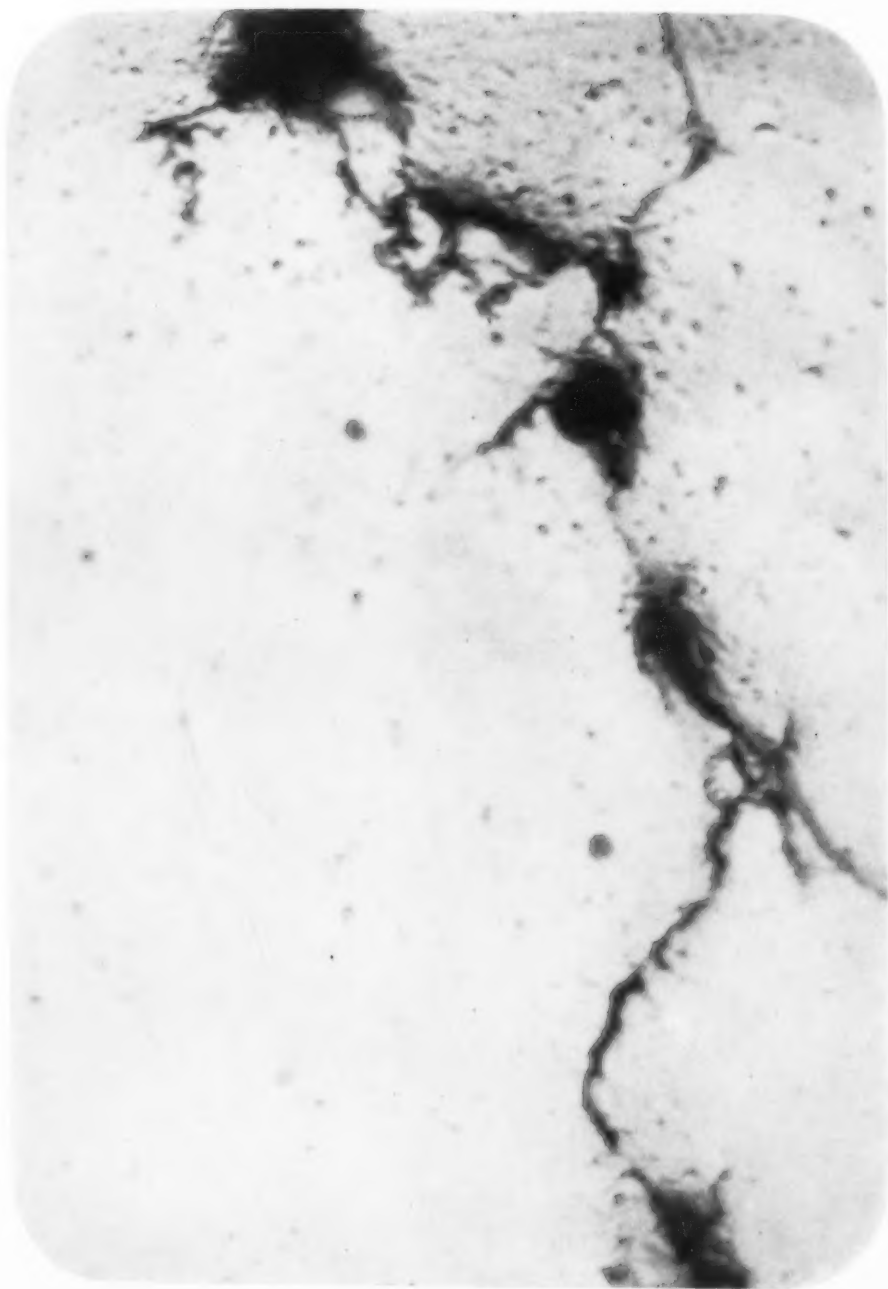


Fig. 8—Magnification 3500X





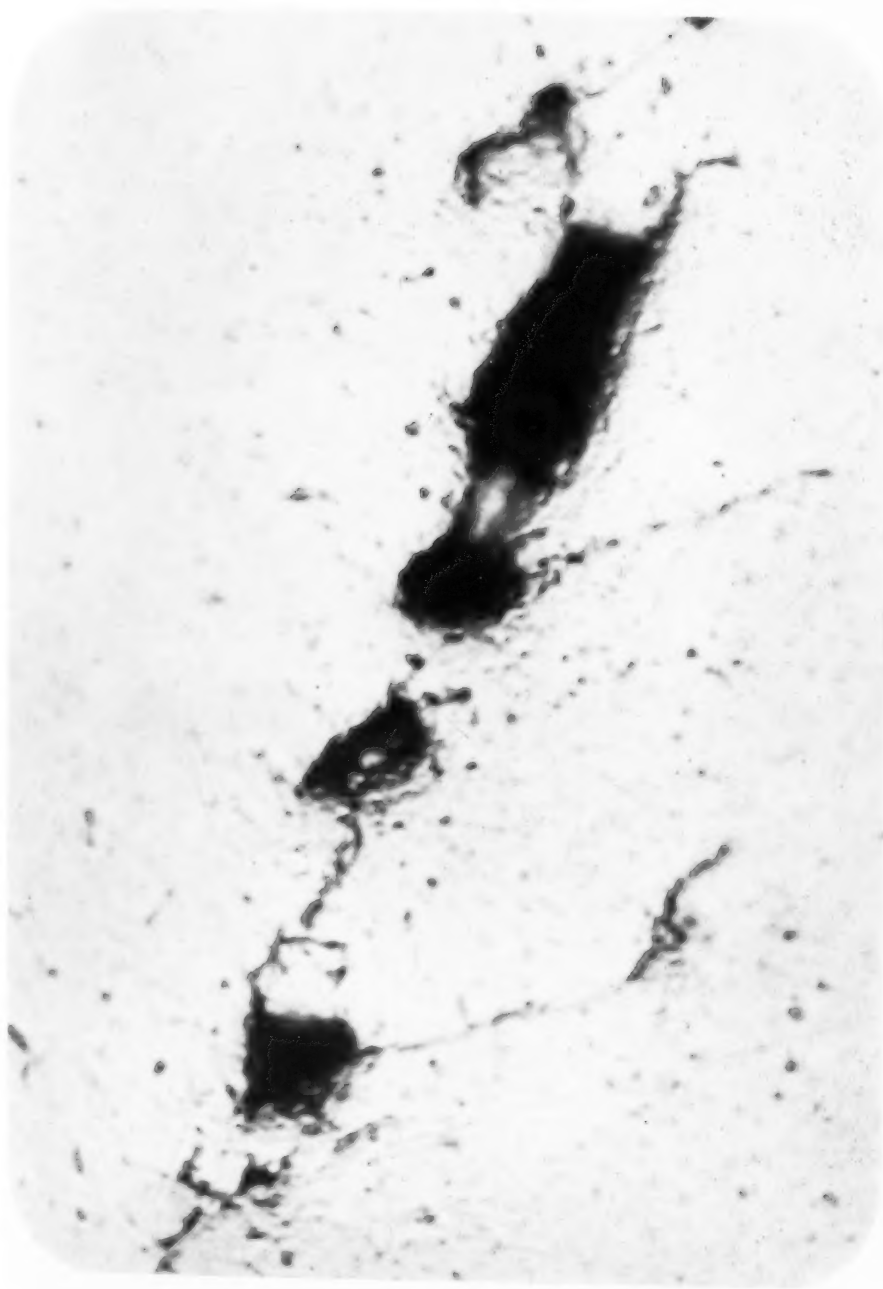


Fig. 9—Magnification 3500X



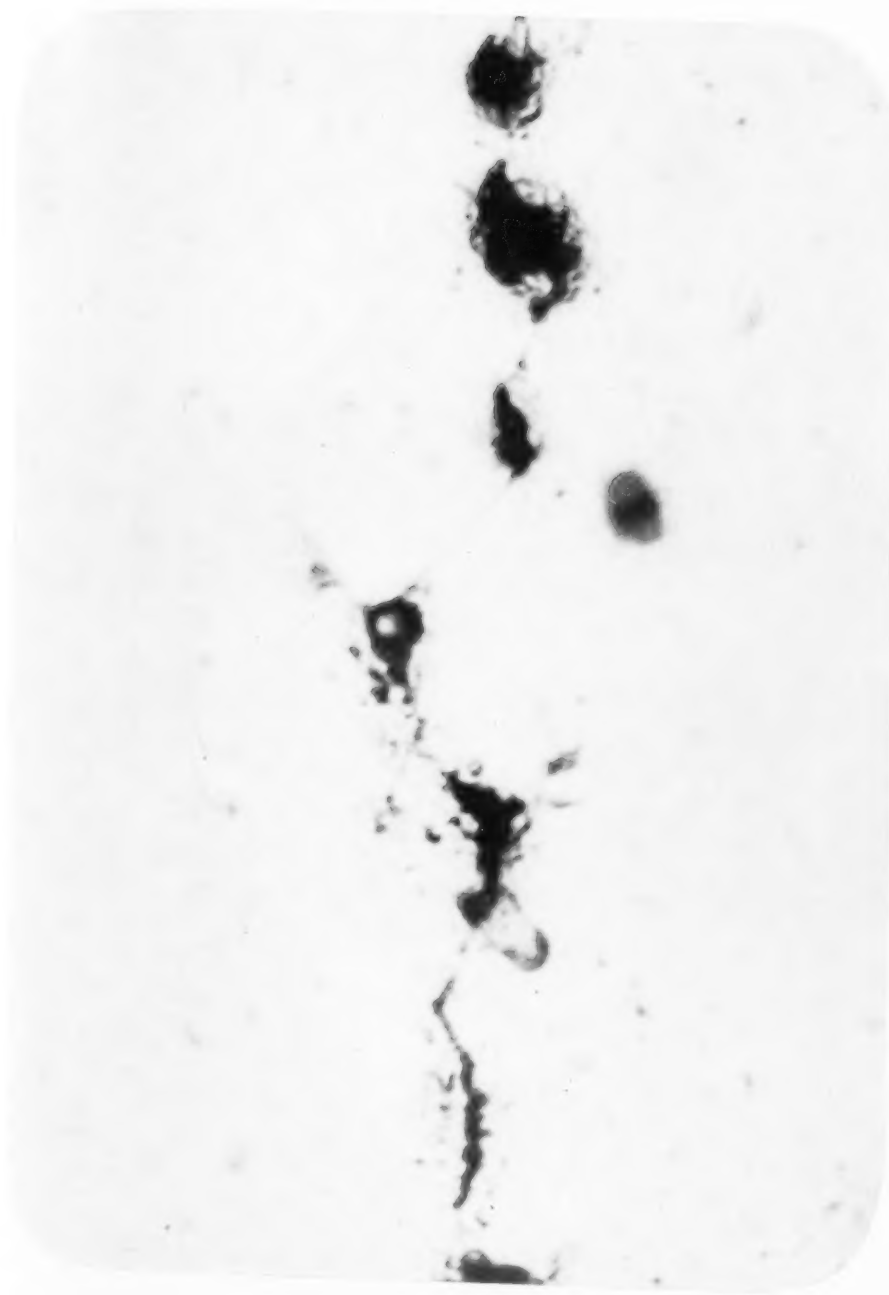


Fig. 10—Magnification 3500X





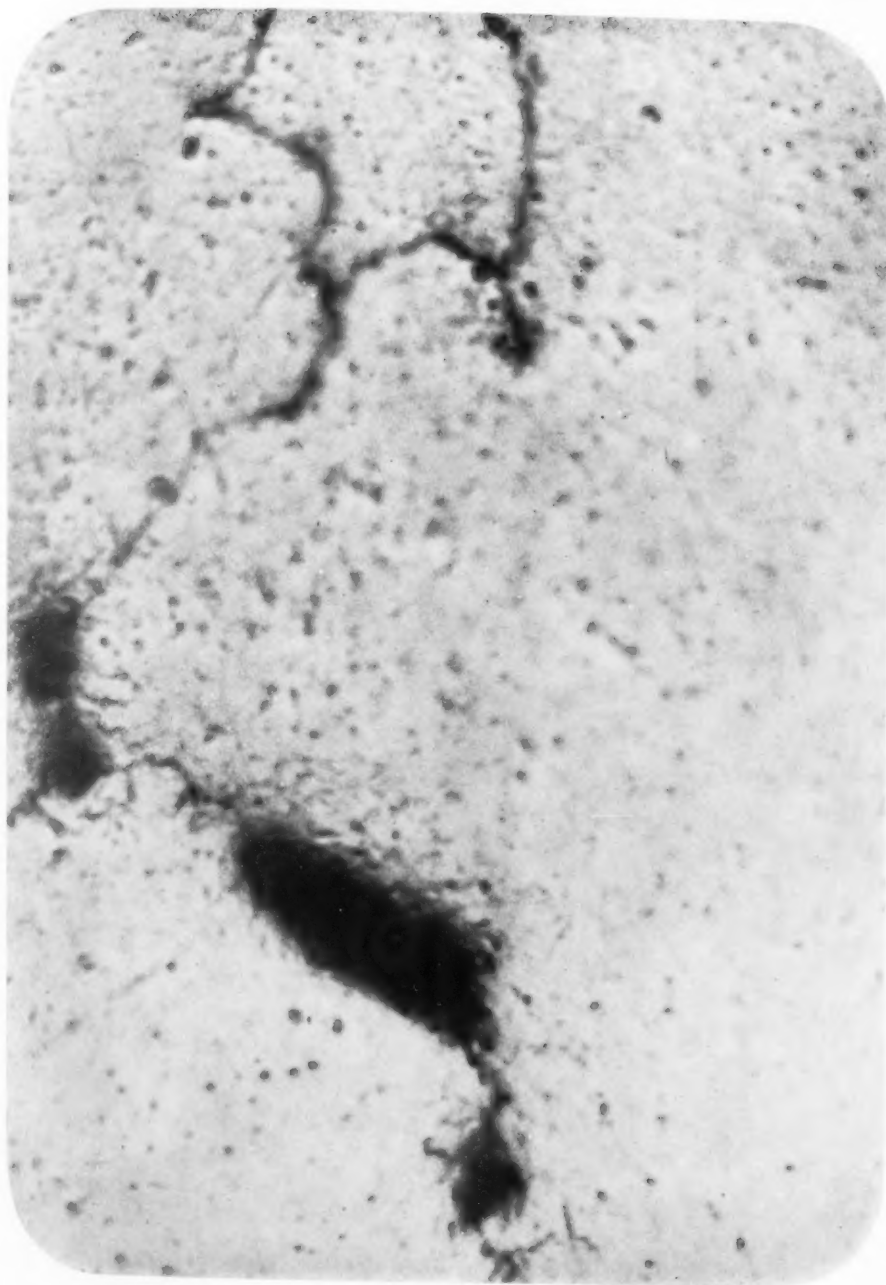


Fig. 11—Magnification 3500X



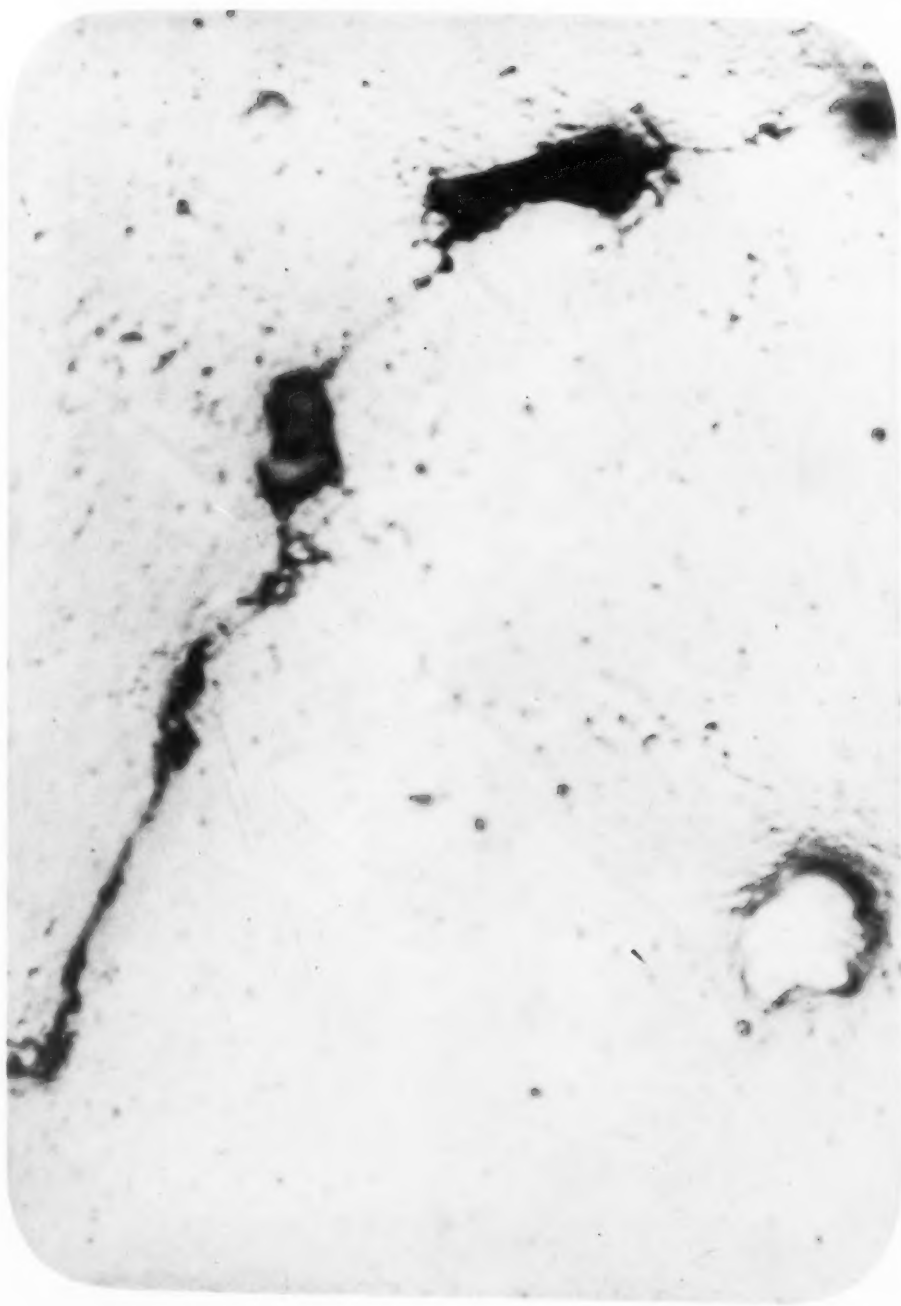


Fig. 12—Magnification 3500X





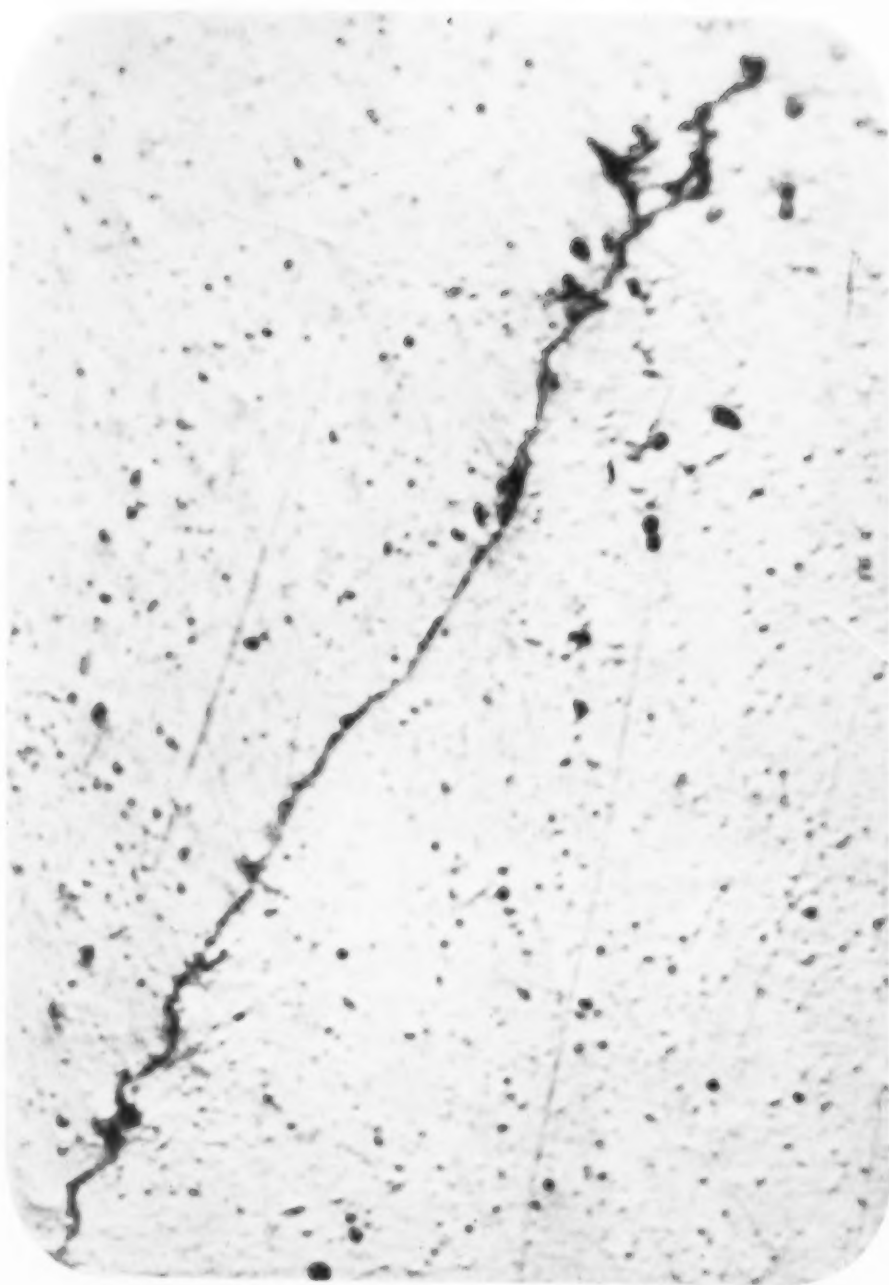


Fig. 13—Apparent end of crack. Magnification 3500X



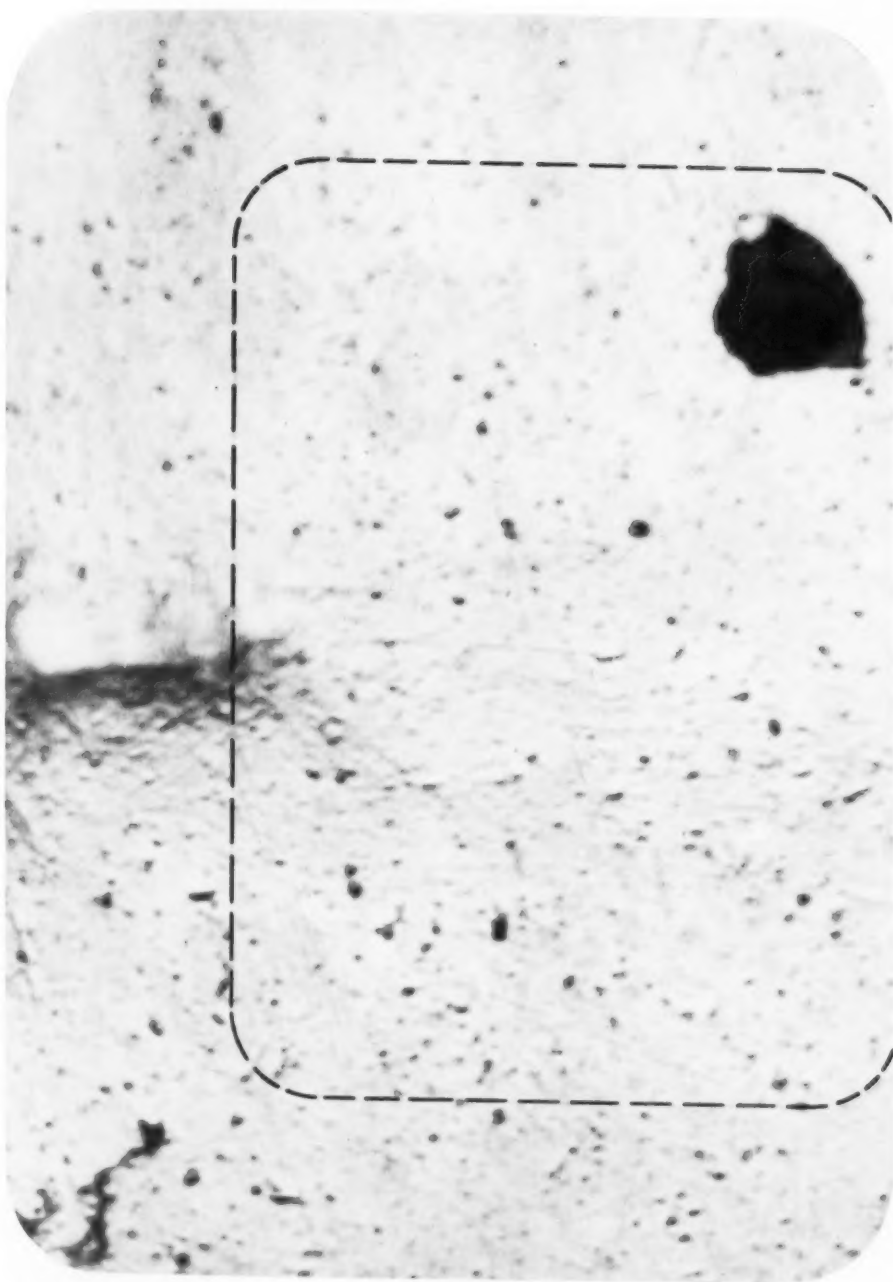


Fig. 14—End of crack trending toward inclusion. Area within dotted lines is shown in Fig. 17.  
Magnification 3500X







Fig. 15—Field immediately below Fig. 14; in upper right is edge of inclusion shown in that figure. Magnification 3500X



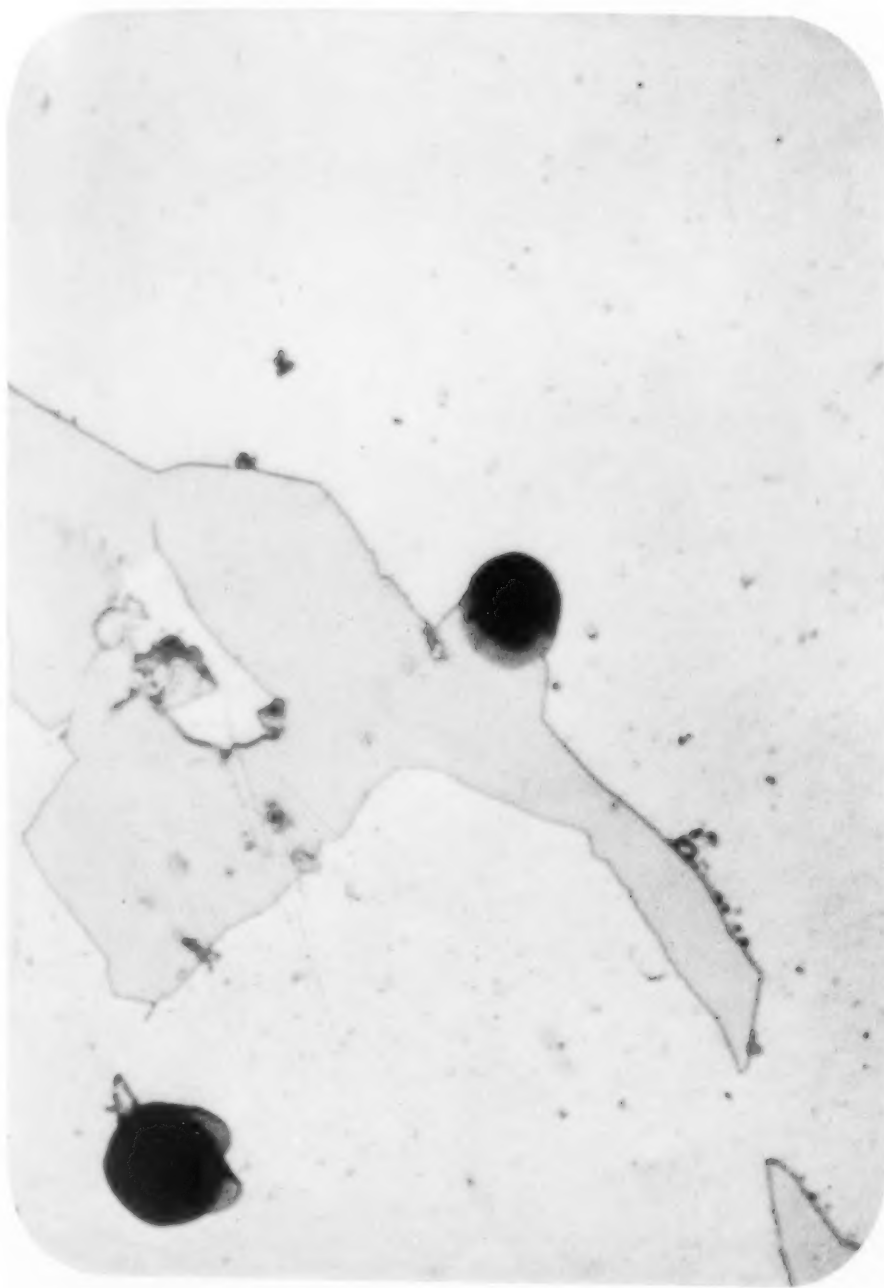


Fig. 16—Typical inclusions occurring in specimen independent of crack. White inclusion, iron carbide; dark inclusions, probably oxides. Magnification 3500X

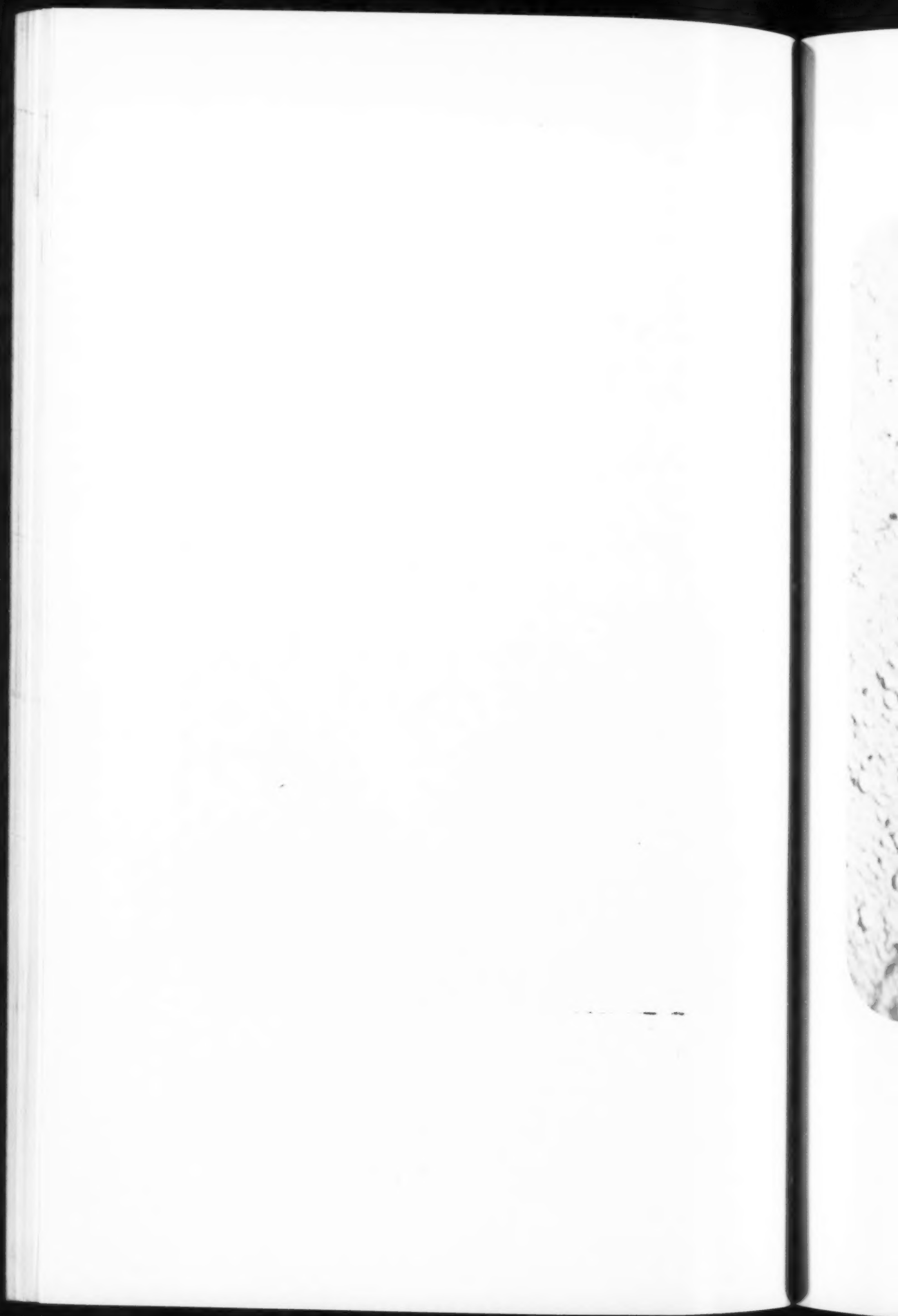






Fig. 17—Area shown within dotted lines on Fig. 14; specimen more deeply etched, crack extends to inclusion. Magnification 4800X



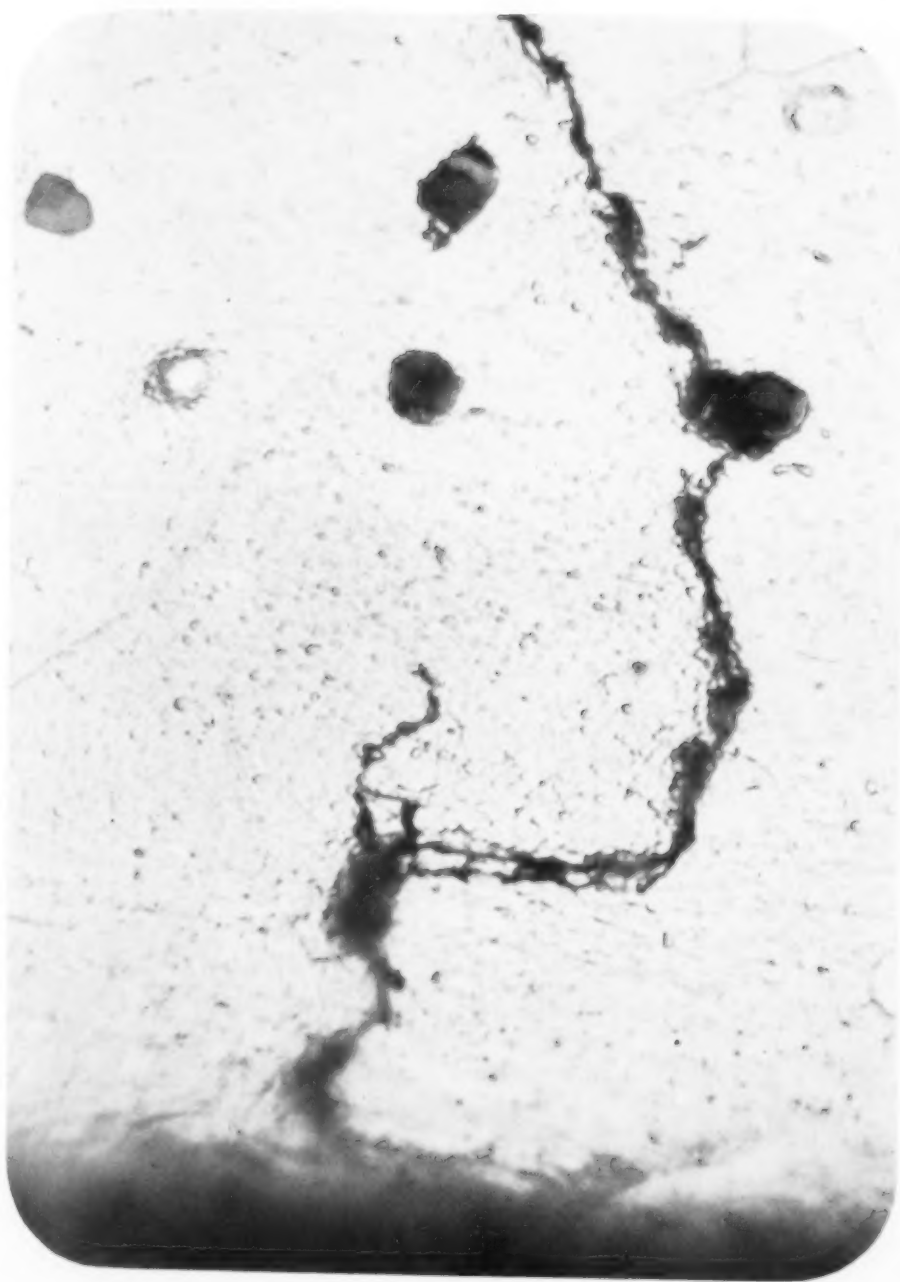


Fig. 18—Origin of crack. Magnification 2170X



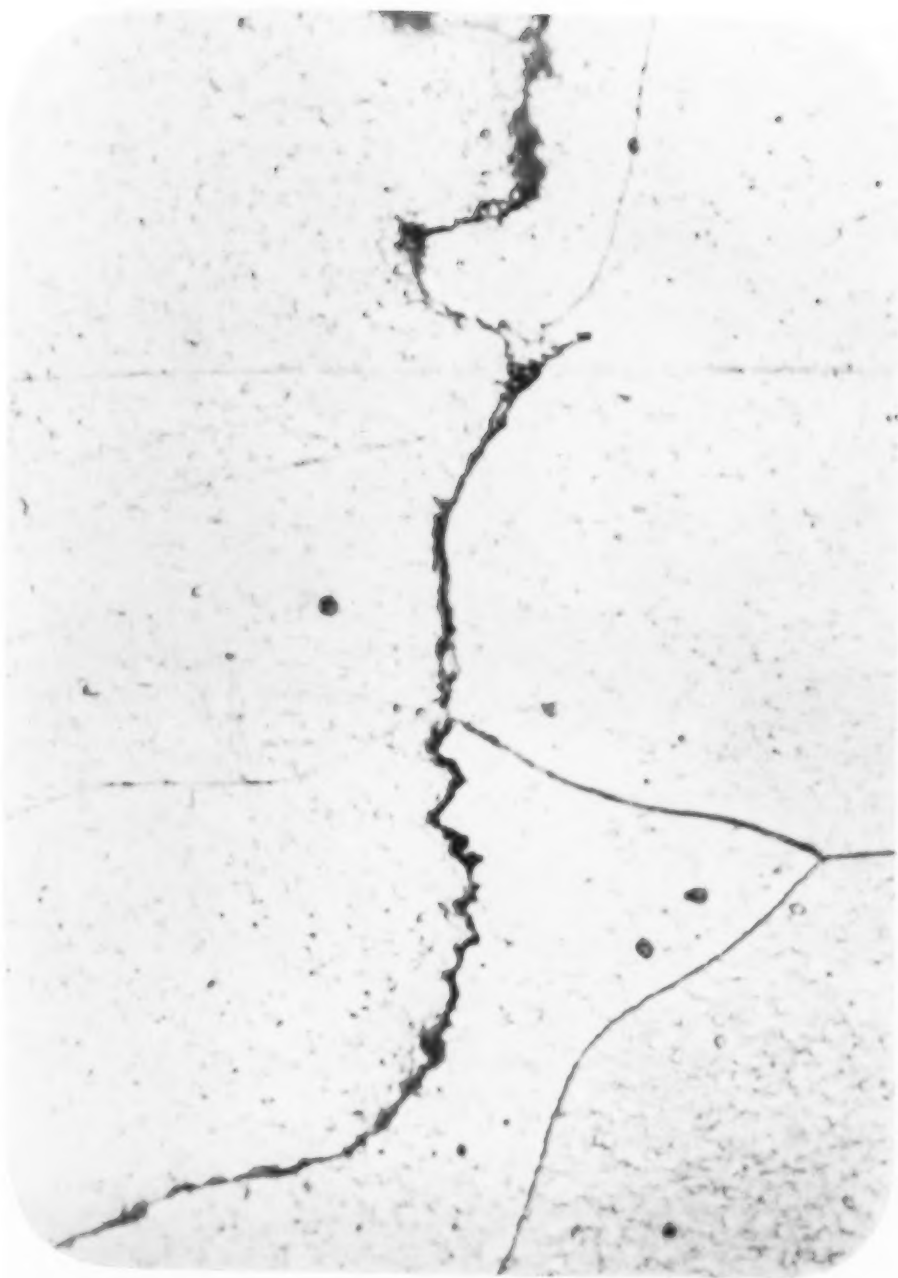
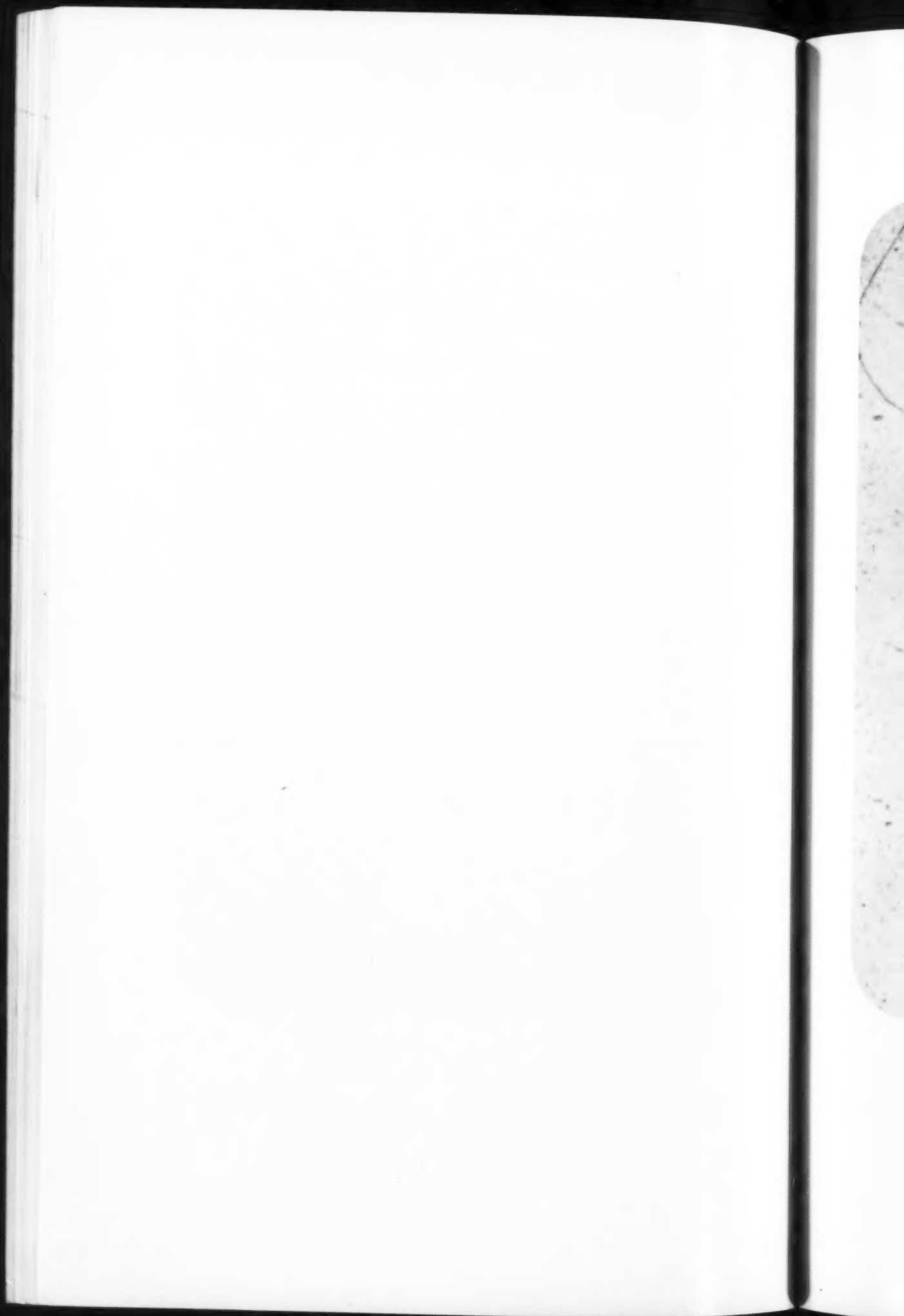


Fig. 19—Magnification 2170X





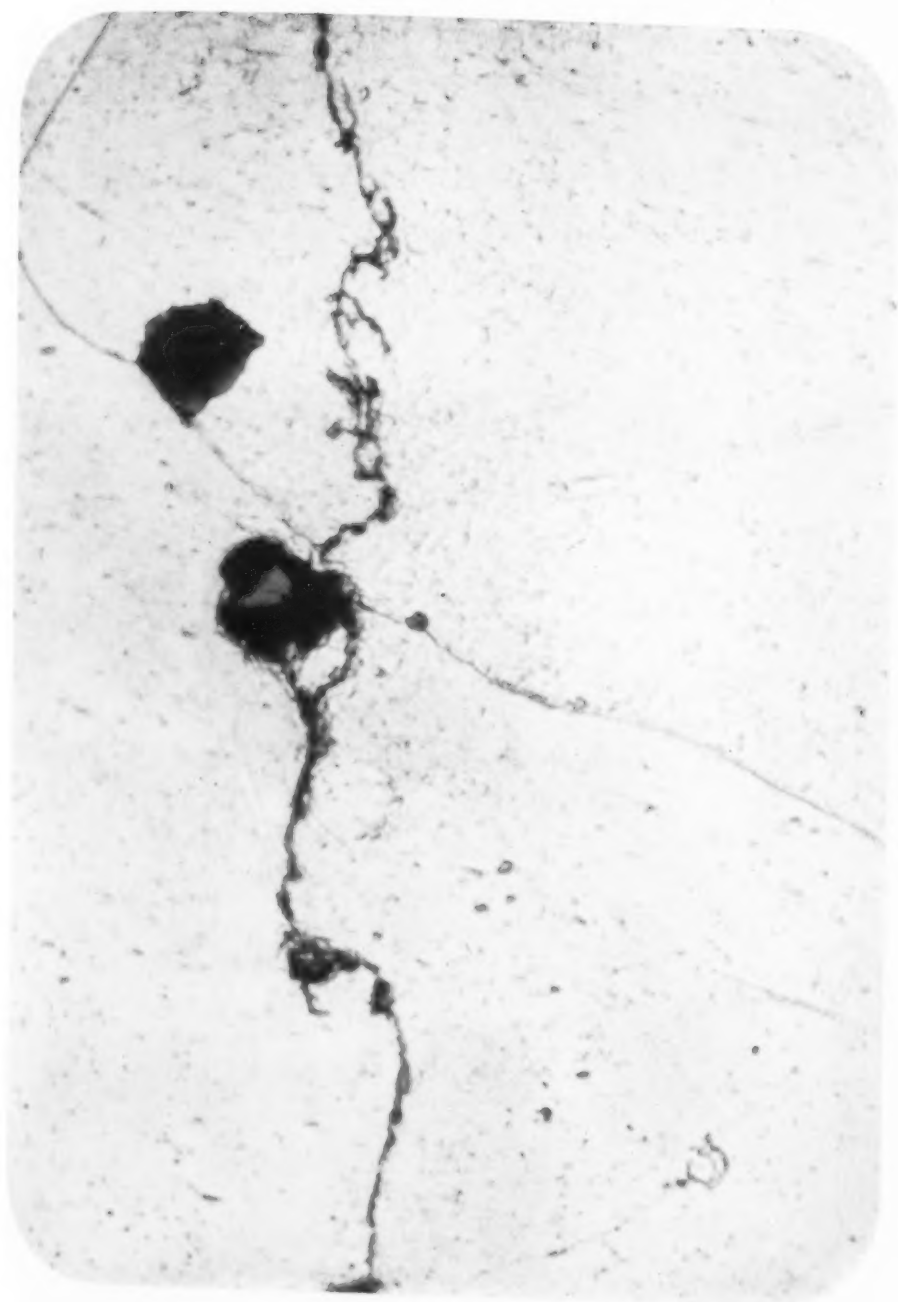


Fig. 20—Magnification 2170X



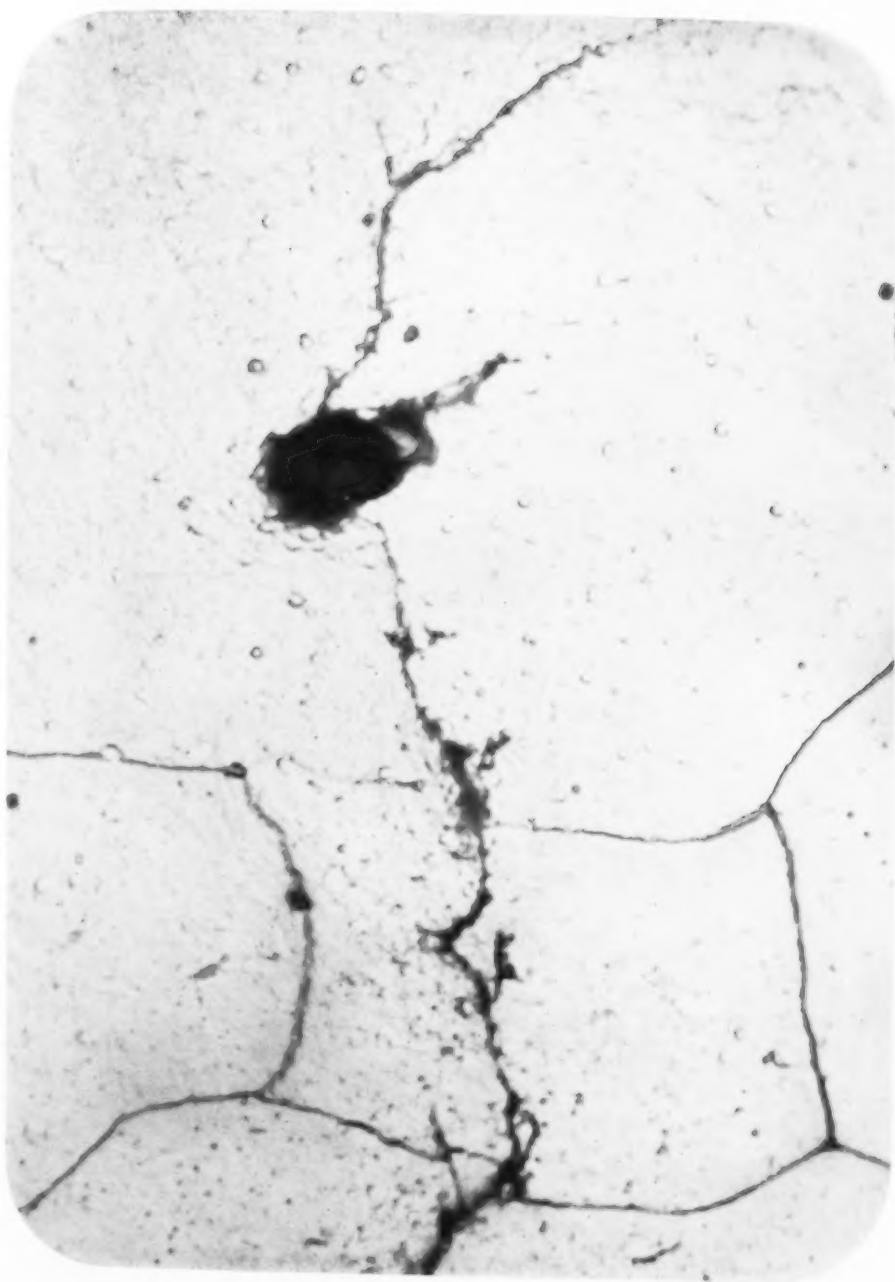


Fig. 21—Magnification 2170X





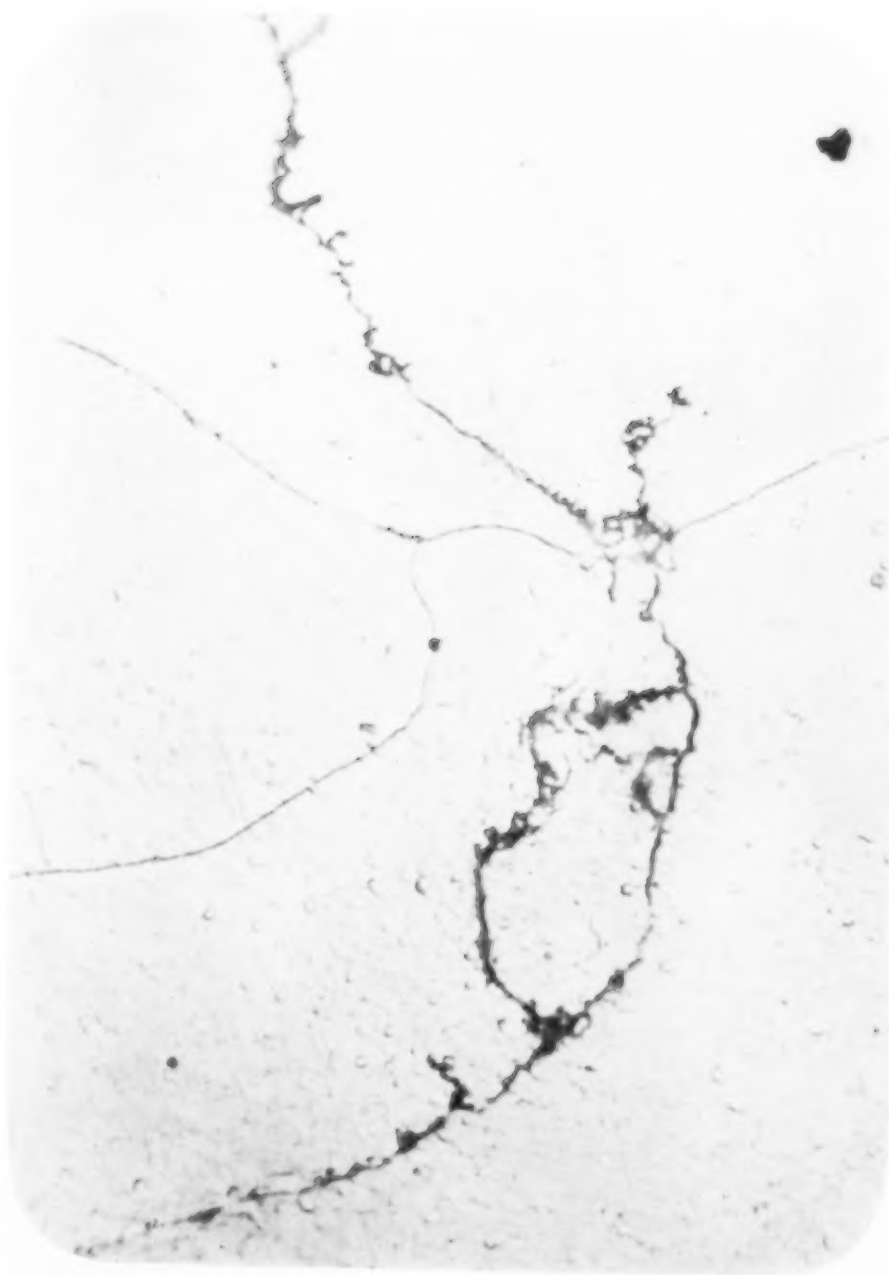


Fig. 22—Magnification 2170X



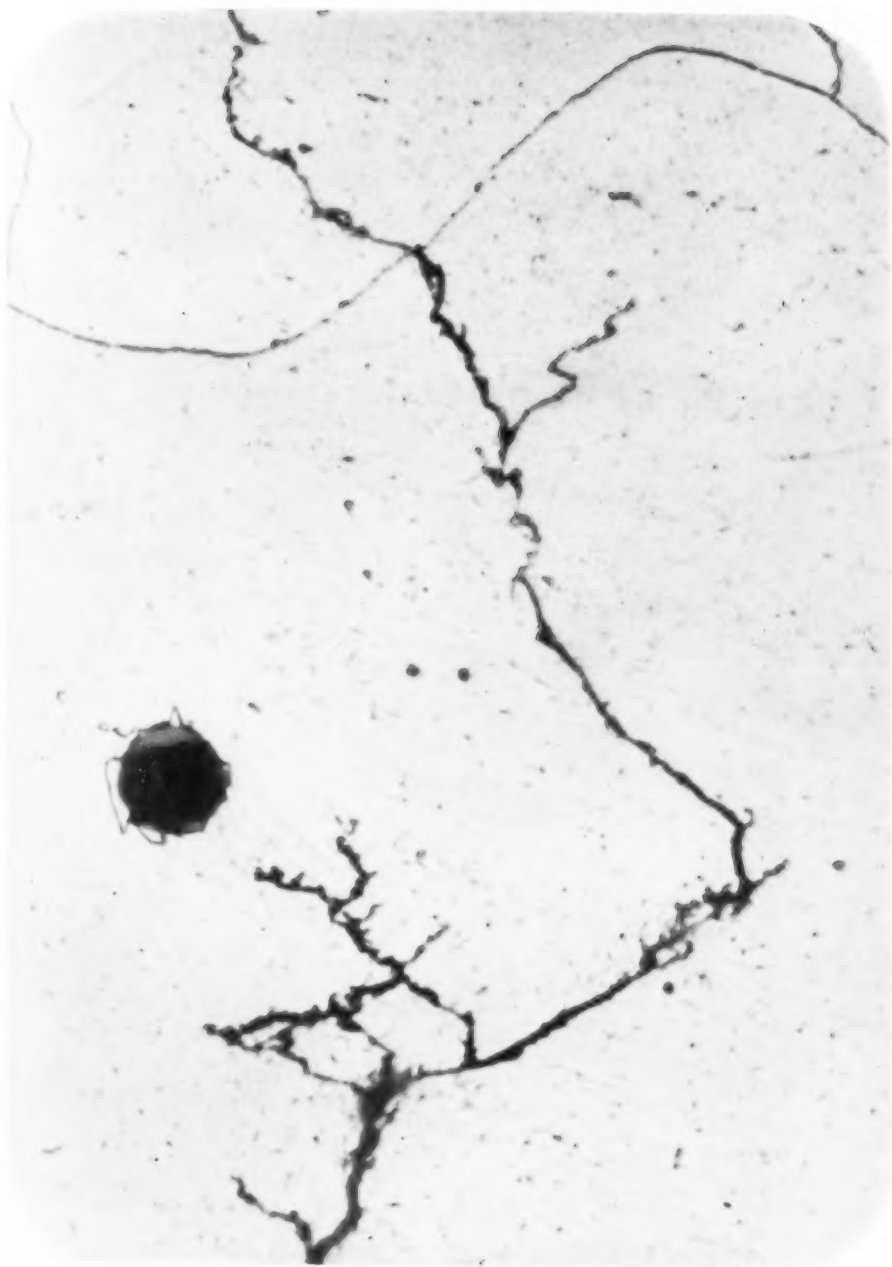


Fig. 23—Magnification 2170X



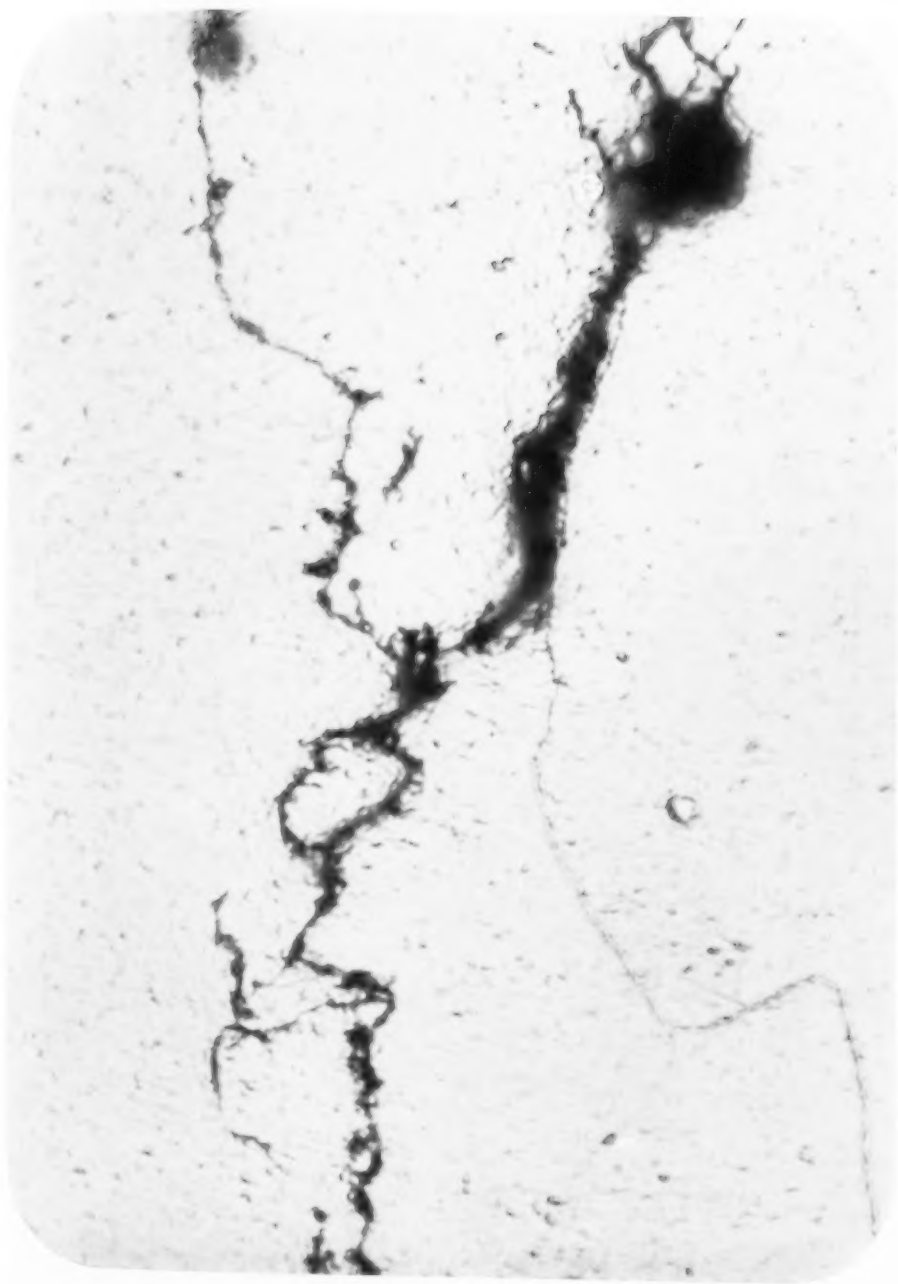


Fig. 24—Magnification 2170X







Fig. 25—Magnification 2170X



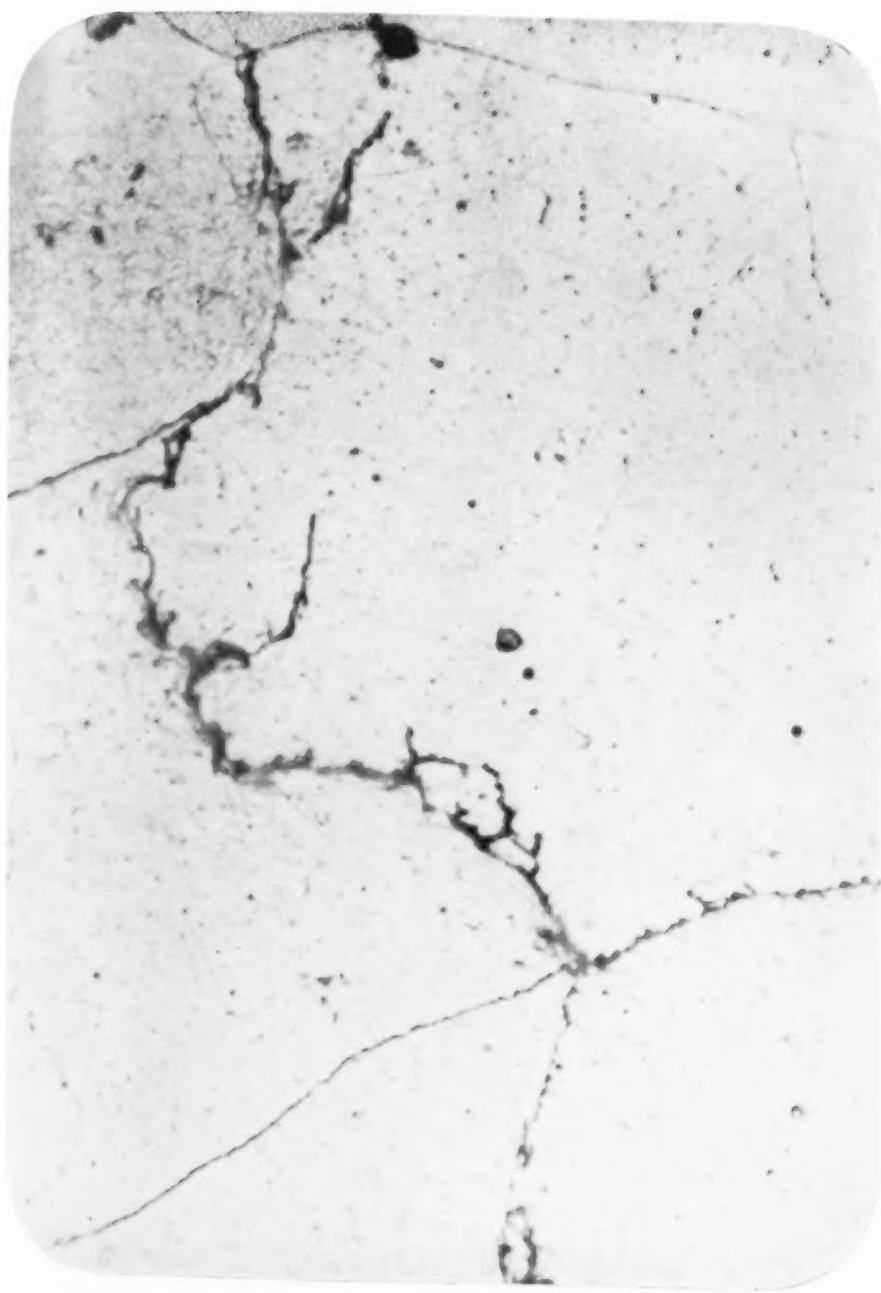


Fig. 26—Magnification 2170X



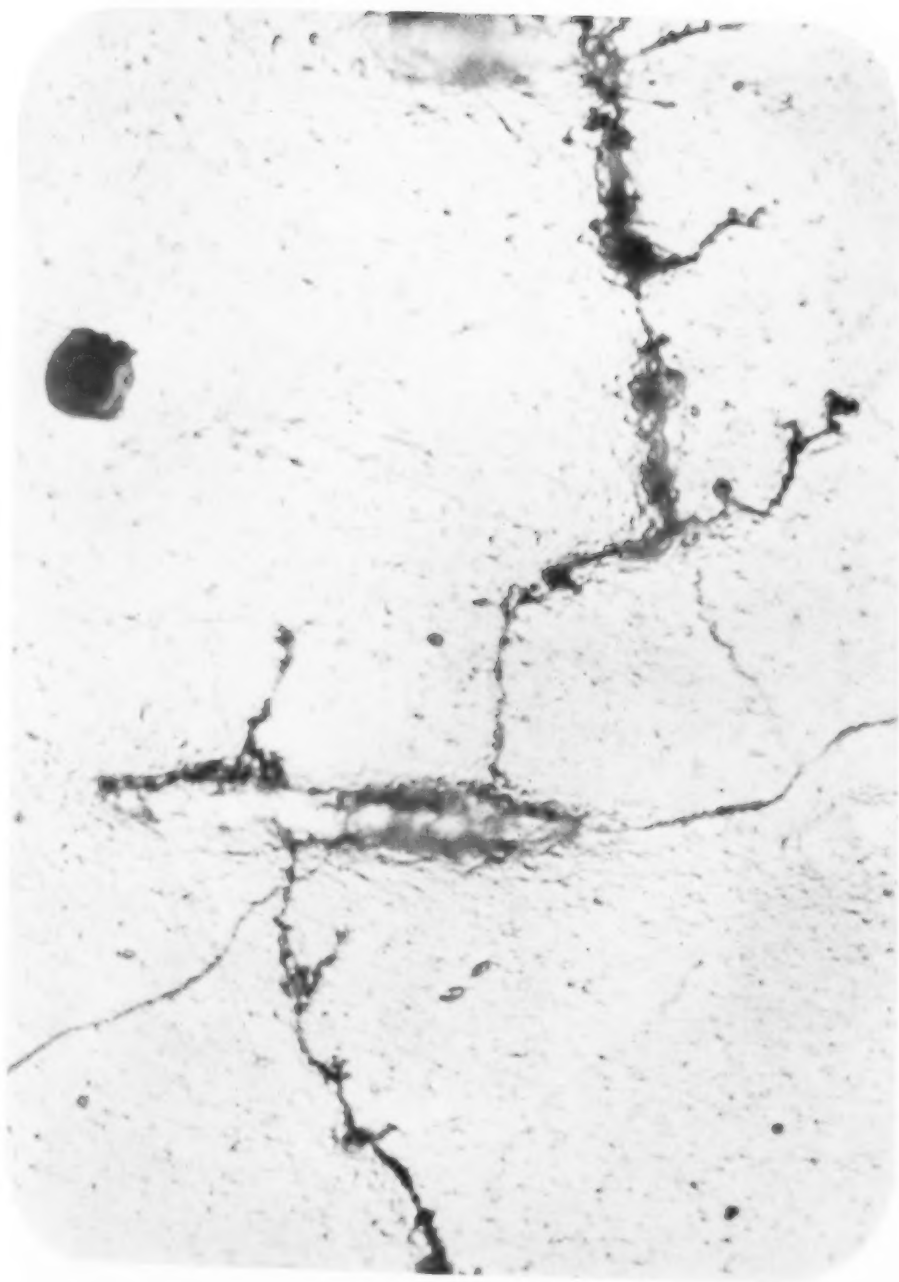


Fig. 27—Magnification 2170X





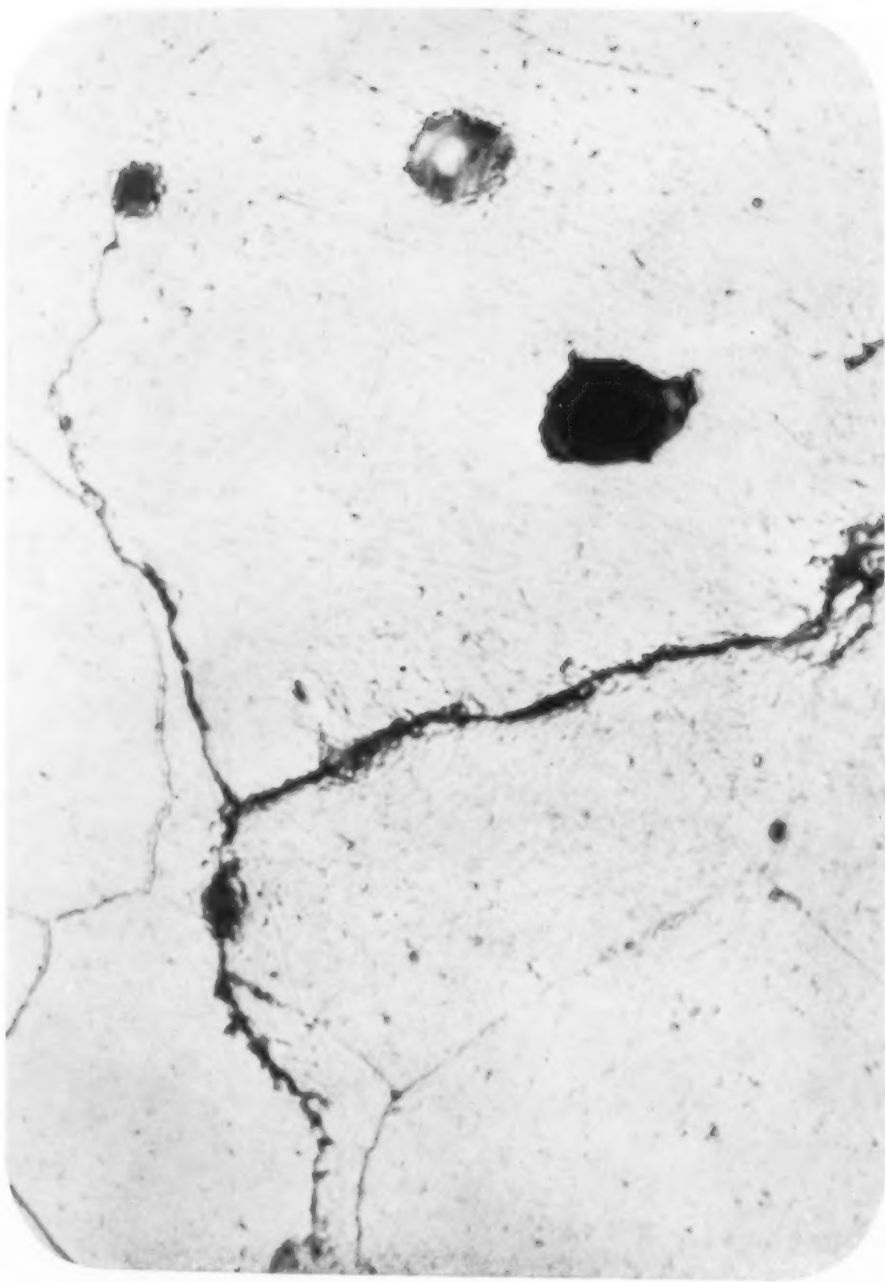


Fig. 28—Magnification 2170X



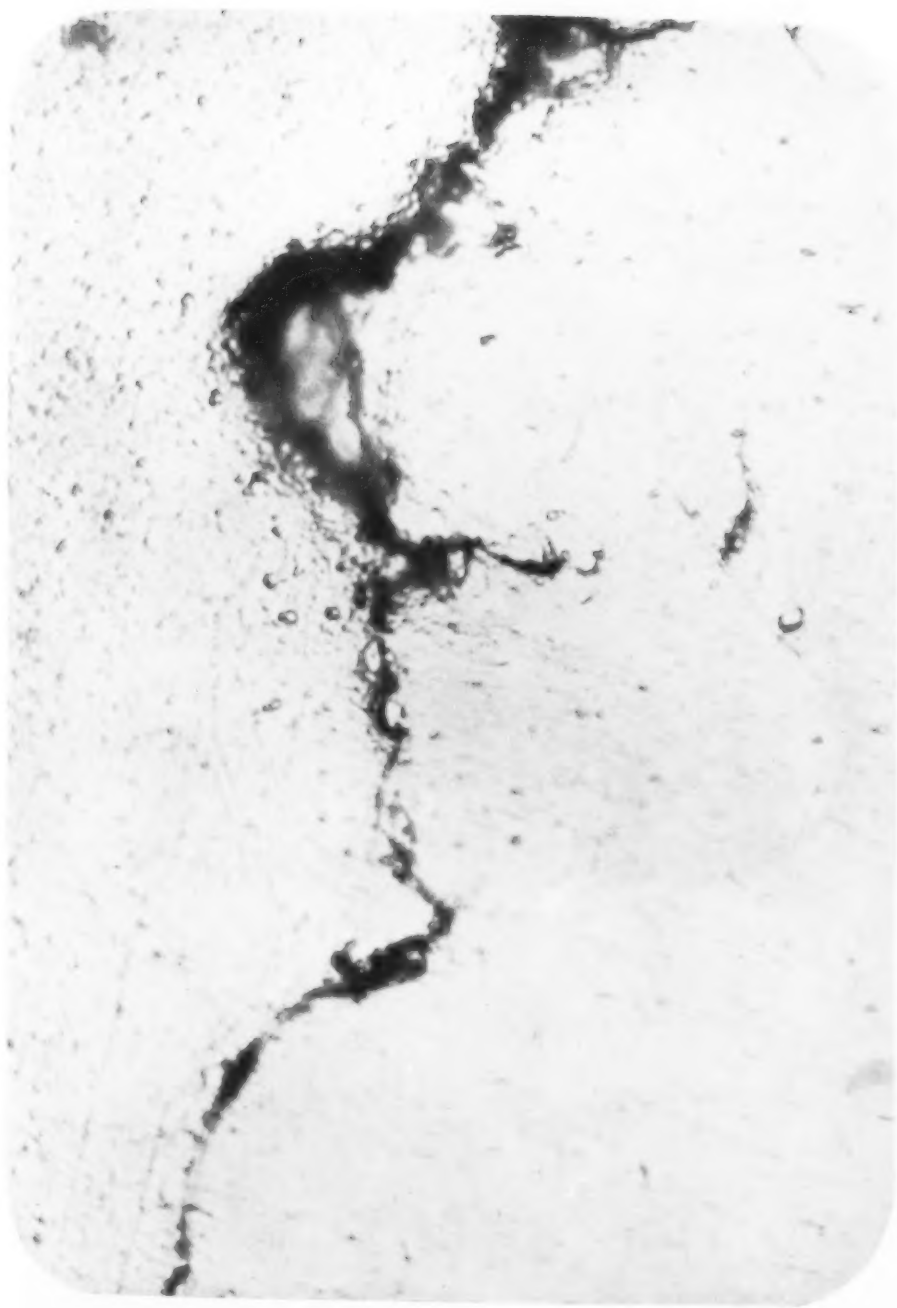


Fig. 29—Magnification 2170X



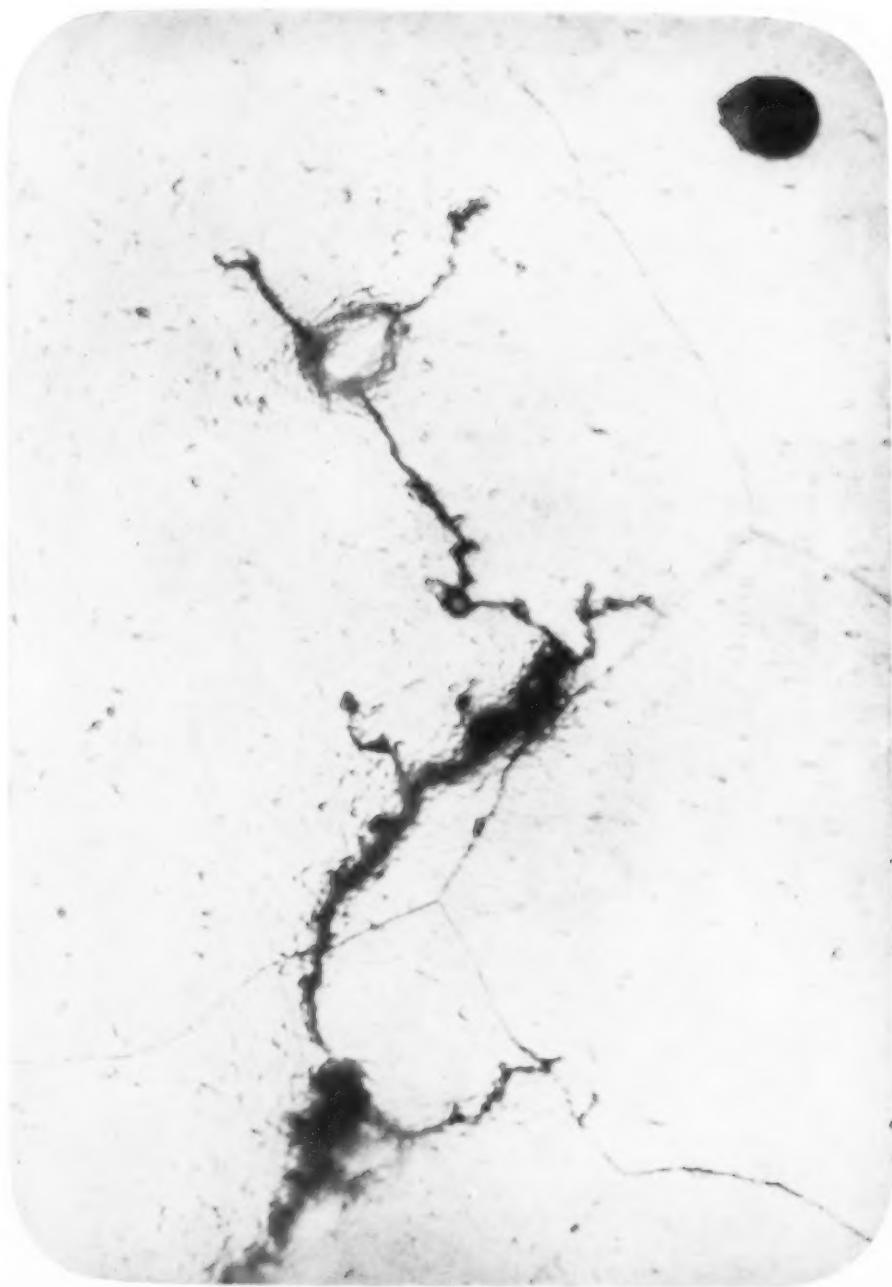


Fig. 30—End of crack. Magnification 2170X





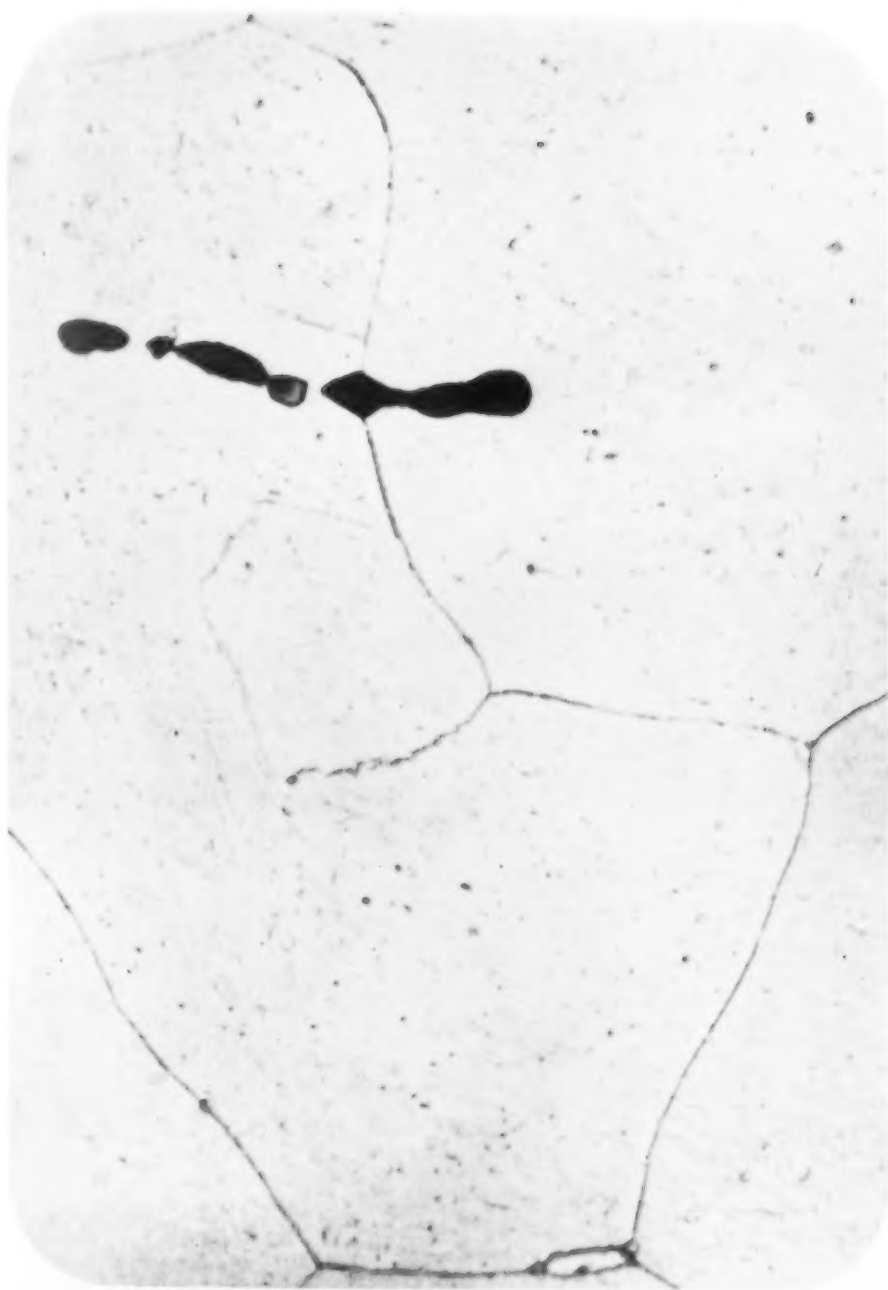


Fig. 31—Small streak of inclusions just beyond end of crack. Magnification 2170X





Fig. 32—Typical inclusions occurring in specimen independent of crack; etched with sodium picrate. Magnification 2170X





Fig. 33—Typical inclusions occurring in specimen independent of crack; etched with sodium picrate. Magnification 2170X



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not give a true estimate of the amount of impurities present in Armeo iron. The assumption, of course, is that many of the inclusions must be oxides. Ferrie oxide, for example, when prepared for microscopic examination appears to have much the same appearance as the slate-blue inclusions found in the Armeo iron.

Visual examination and the photographic results show that the Armeo iron contains innumerable inclusions of small magnitude but nevertheless the included matter seems to have a decided bearing on the ability of the metal to withstand fatigue.

In a recent paper<sup>3</sup> on "Corrosion-Fatigue of Metals", McAdam describes the results obtained with certain steel specimens tested in a water stream. He states:

"As the specimens so tested were entirely surrounded by the water stream they had not become rusty at the end of this time interval ( $1\frac{1}{4}$  to 4 hours) but appeared fairly bright. On closer examination, however, small spots of dull appearance were visible without a hand lens. When these spots were examined at higher magnification the cause of the damage to the endurance properties was apparent.

—Some of these spots are oxide patches each surrounding a non-metallic inclusion or pit from which an inclusion has been removed,—some of the spots show no decided oxide patches but consist of inclusions (or pits from which inclusions have been removed) from which cracks have started. The cracks are all nearly transverse to the axis of the specimen."

The inclusions described by McAdam may be of an order of magnitude greater than those dealt with in this paper. He concludes that non-metallic inclusions have a harmful effect on the endurance properties of iron and steel.

During the work on the fatigue specimens a secondary structure was developed at high powers irrespective of the location of the fatigue crack. This secondary structure appears to be an inherent characteristic of Armeo iron and probably results from preparation or fabrication of the metal. Since it seems not to be tied up definitely with the fatigue resistance characteristics of the metal, it will not be dealt with at this time.

<sup>3</sup>"Stress-Strain-Cycle Relationship and Corrosion Fatigue of Metals," D. J. McAdam, Jr. Paper presented before the 29th Annual Meeting A. S. T. M., June, 1926.

### SUMMARY

The results of the investigation to date may be summarized as follows:

1. Non-metallic inclusions in Armco iron are a potential source of weakness when such metal is subjected to reversed cycles of stress.
2. Iron carbide when occurring as inclusions in Armco iron is not a source of weakness.
3. Non-metallic inclusions seem to be insecurely seated in the metal and the boundary between the inclusion and the metal is the path often followed by the fatigue crack.
4. The non-metallic inclusions act as "stepping-stones" for the fatigue crack. (A description suggested to the writer by Dr. J. A. Mathews and adopted as a most appropriate analogy.)
5. Grain boundaries do not appear to be a potential source of weakness.
6. Reversed cycles of stress appear to produce disturbances in the structure of the metal in advance of the visible crack, as disclosed by light etching. This condition at times can be disclosed by deeper etching.
7. Disturbances in the metal structure adjoining the path of fatigue failure seem to be highly localized to the immediate neighborhood of the crack.
8. Slip planes, strain lines, Neumann bands or other similar markings were not found if the strained condition of the metal immediately adjoining the crack is disregarded.

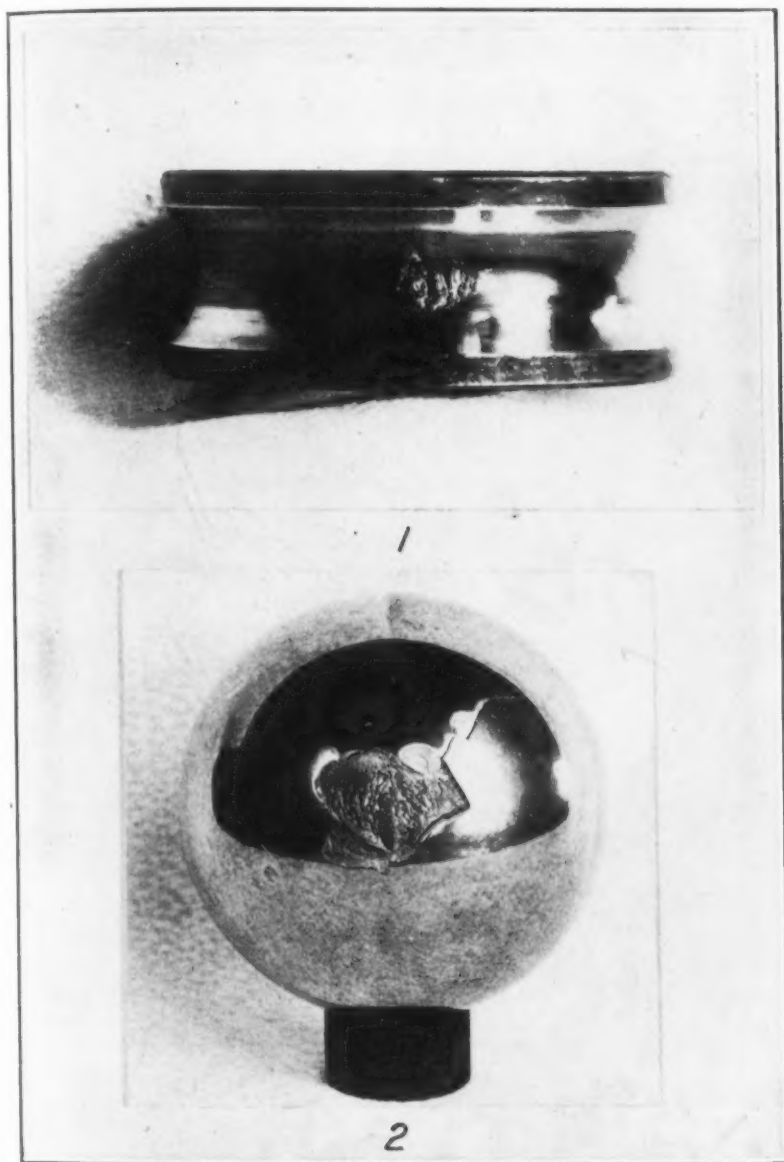
### Oral Discussion

J. M. LESSELLS: There are one or two points which should be mentioned. The first is I would like to draw to Mr. Lucas' attention that from the specimen he showed it seems to me that the path of fatigue has been more or less predetermined. In other words, he uses a specimen where there is a slight stress concentration, and I would like to ask him if this will affect the results in any way.

The second point I would like to make is that it is to be hoped that Mr. Lucas will extend his work to other steels. There has been a certain amount of other work already done, and I refer to that published by H. J. Gough of the National Physical Laboratory in England, but it is to be hoped that he will extend his work to, say, heat treated steel. In the case of ARMCO iron, we know that the endurance limit—I think this point was stressed by Professor H. F. Moore—is much higher than the elastic limit in tension, whereas in heat treated steel, say nickel-chromium steel, the endurance limit is much

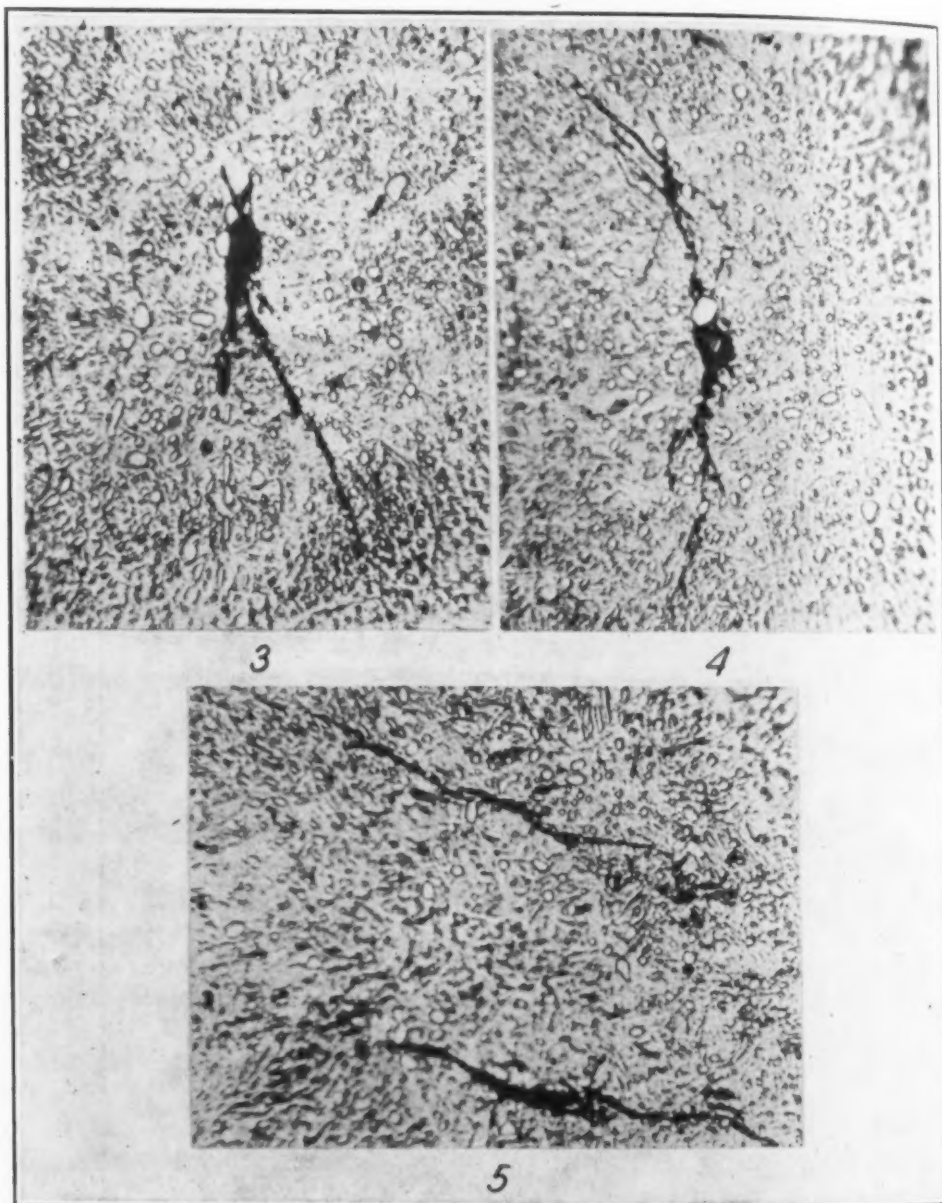
lower, so that for such a material you may get a proportional limit of 80,000 pounds per square inch and an endurance limit of 50,000. Moreover, in ARMCO iron I think I am right in saying it is comparatively easy to study the path of fatigue, but in those heat treated steels, where the crystals are much more finely divided, it is a much more difficult problem. To my mind this high power magnification might throw some light on this phenomenon.

DR. H. STYRI: It is with extreme pleasure that I have had the opportunity to listen to this talk and the opportunity to view these fine pictures, and I wonder if you all realize the importance of impurities like slag on the



Figs. 1 and 2—Illustrating Failures of an Inner Bearing Race and a Ball.

fatigue resistance of steels. In our work on fatigue of ball bearings we have for years been aware of the great danger of masses of impurities and we have tried to emphasize on our sources of steel supply, how very important it is to have clean material.



Figs. 3, 4 and 5—Photomicrographs of Chromium Bearing Steel Showing Slag Inclusions and the Beginnings of Fatigue Cracks. Mag. 100x.

To illustrate, I would like to show where fatigue in ball bearings happens and by means of some pictures add something to Mr. Lucas' talk. Characteristic fatigue failures are shown in Figs. 1 and 2, illustrating failures of an

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inner race and of a ball. In sixty per cent of the cases of ball failures, I would say, we can trace them to slag inclusions. In the inner race this cause is difficult to find, and likewise in the outer race, because of the nature of the failure, but when we take a cross-section of a race after certain test conditions, we find the slag material is disturbed, and by careful polishing and etching find a characteristic structure much resembling streaky troostite. Sometimes the streaks are sharp and well developed and we have found them even if the piece has not failed. The photomicrographs shown in Figs. 3, 4 and 5 have been taken just below the surface on races which have not failed, where we could not find any sign of failure on the surface.

The material is chromium bearing steel of a Brinell hardness of about 640. The slag inclusion is about 0.0004 of an inch in diameter, and the crack is the beginning of a fatigue crack. (Fig. 3.) These pictures were taken by Mr. Walp in our laboratory at 1000 diameters magnification. The second one shows starting fatigue cracks and the slag inclusion is about 0.0002 of an inch in diameter. (Fig. 4.) The third one shows the beginning of fatigue cracks where we did not happen to strike the slag inclusion. (Fig. 5.)

CHAIRMAN JEFFRIES: If there is no further discussion, before Mr. Lucas closes I would like to call attention to one point, and that is that the non-metallic inclusions in soft material like ARMCO iron actually raise the elastic limit and tensile strength, so that it is especially interesting to find that such inclusions lower the fatigue resisting properties of one of these materials. Of course, that only goes to show what we have suspected for a long time, that the mechanisms of fracture must be inherently different, inasmuch as the fracture of fatigue takes place without reduction of area and the fracture in tensile test of one of these soft materials takes place with marked reduction of area, and the function of the nonmetallic inclusion in the tensile test is obviously simply to act as a resistance to slip along the slip planes of the crystals, and they function effectively in that capacity, because they are harder than the material in which they are imbedded. The fact, however, that such non-metallic inclusions do not have any considerable adhesion to the metal would certainly be favorable for the fatigue fracture, which takes place without substantial general deformation and without reduction of area.

**Written Discussion:** By A. V. de Forest, metallurgist, American Chain Company, Bridgeport, Conn.

As usual, the extremely clear photographs taken by Mr. Lucas give us a new knowledge of whatever phenomena he chooses to study. In this case he supplies long needed evidence that fatigue failures follow a particular type of inclusion. It is greatly to be hoped that more work along the same line will settle the interesting question as to what happens when there are no inclusions for the crack to follow.

A further point of great importance in understanding the mechanism of fatigue is the question of what changes in stress or structure precede the formation of the crack. In other words, does a certain repeated deformation produce a permanent alteration in the general volume of the metal or only over a microscopically thin section which is to be occupied by the crack. Does this changed portion reach out and locate the inclusions, as appears possible from the photographs, or does the crack continue its course haphazard as



directed by the metal in its immediate path? As Mr. Lucas indicates the inclusions themselves may have effects at a distance which warn the crack of their presence, some attracting and others repelling its advancing edge.

In the present case there seems to be no change in the whole volume of metal, which presumably has suffered the same stress, but perhaps not the same strain as that in the path of failure. More work is urgently needed to decide this point, for it is of considerable practical importance. For instance, if it be established that concentrations of strain are necessary as a preliminary to crack formation, there should be an improvement in fatigue life by moderate heat treatments at suitable intervals. If, on the other hand, the energy supplied to the test specimen is at once utilized in forming the crack, a redistribution of stress is of no great value.

This paper seems to me especially valuable in showing that with skillful handling, high power microscopy can become a valuable tool to the engineer, as it has already become the main stay of the metallographer.

**Written Discussion:** By H. F. Moore, research Professor of Engineering Materials, University of Illinois; in charge, Investigation of the Fatigue of Metals.

Mr. Lucas' paper is a fruitful excursion into one of the most promising regions of the field of the fatigue of metals. His paper not only describes clearly the phenomena observed, but emphasizes in a vivid way the importance of minute inclusions as "stepping stones" for fatigue failure. I believe that his work shows more clearly even than the previous work of metallographists how our conventional concepts of stress and strain must be modified when we consider fatigue failure of metals. We have thought of metal as homogeneous and capable of indefinite division without change of properties. As regards elastic strength,—the determination of some value called an "elastic limit"—this is a practical working concept because a considerable mass of metal must be distorted before inelastic action can be detected, but in considering fatigue strength this concept is not altogether satisfactory because the progressive spread of a fatigue crack in its early stages involves a very small mass of metal.

Mr. Lucas' paper suggests strongly the reasonableness of A. A. Griffith's\* theory of failure of metals. This theory is: that scattered throughout the mass of a piece of metal there are vast numbers of minute flaws or discontinuities quite regularly distributed. These flaws are for the most part even smaller than those shown by Mr. Lucas. The intensified stress at these discontinuities accounts for the great difference between the measured tensile strength or shearing strength and the theoretical cohesion of the metal, which is much higher.

The speaker considers the American Society for Steel Treating very fortunate in receiving this paper of Mr. Lucas and the National Research Committee on Fatigue Phenomena very fortunate in securing the cooperation of Bell Telephone Laboratories and of Mr. Lucas.

**Written Discussion:** By J. Muller, research engineer research department, A. O. Smith Corp., Milwaukee, Wisconsin.

\*Phenomena of Rupture and Flow in Solids," *Philosophical Transactions*, Royal Society, A, Vol. 221, p. 163 (1920).

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The paper presented by Mr. Lucas is indeed a valuable contribution toward the answer to the question of fatigue of metals.

The writer agrees with the author that numerous cycles of stress in fatigue are likely to break the bonds and loosen the inclusions which are directly in the path of the crack. It does not seem likely, however, that

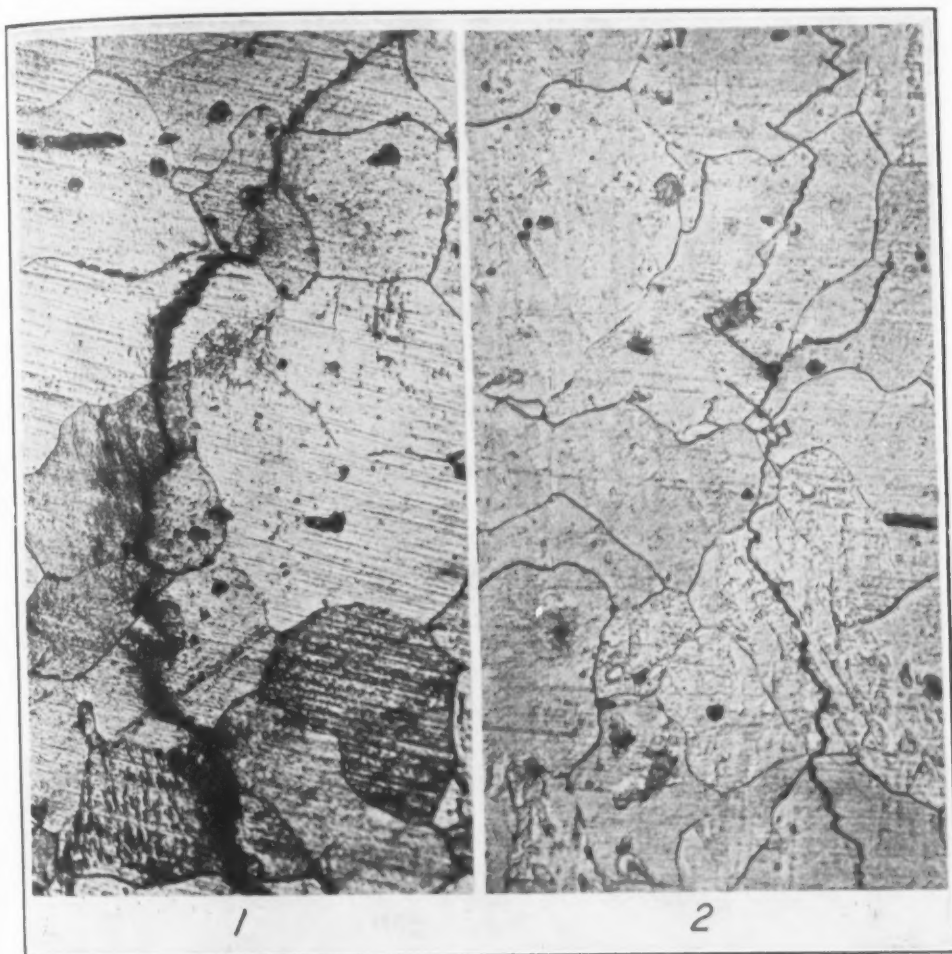


Fig. 1—Fatigue Crack at Surface of Armco Iron. Thickness of Specimen 0.100-Inch. Mag. 350x. Fig. 2—Fatigue Crack 0.011-Inch from Original Surface of Armco Iron. Mag. 350x.

inclusions are the controlling factor in directing the path of the fatigue fracture through the material.

The writer believes that the author's high magnification photomicrographs have obscured an aspect of the fatigue question which is obtained at lower magnification.

The writer finds that in lower magnification photomicrographs the crack does not include inclusions which it may be properly expected to include. The crack as seen at this magnification appears jagged and does not seem to particularly seek inclusions. The abrupt change of direction of the crack



Fig. 3—Fatigue Crack at 0.016-Inch from Original Surface of Armco Iron. Mag. 350x.

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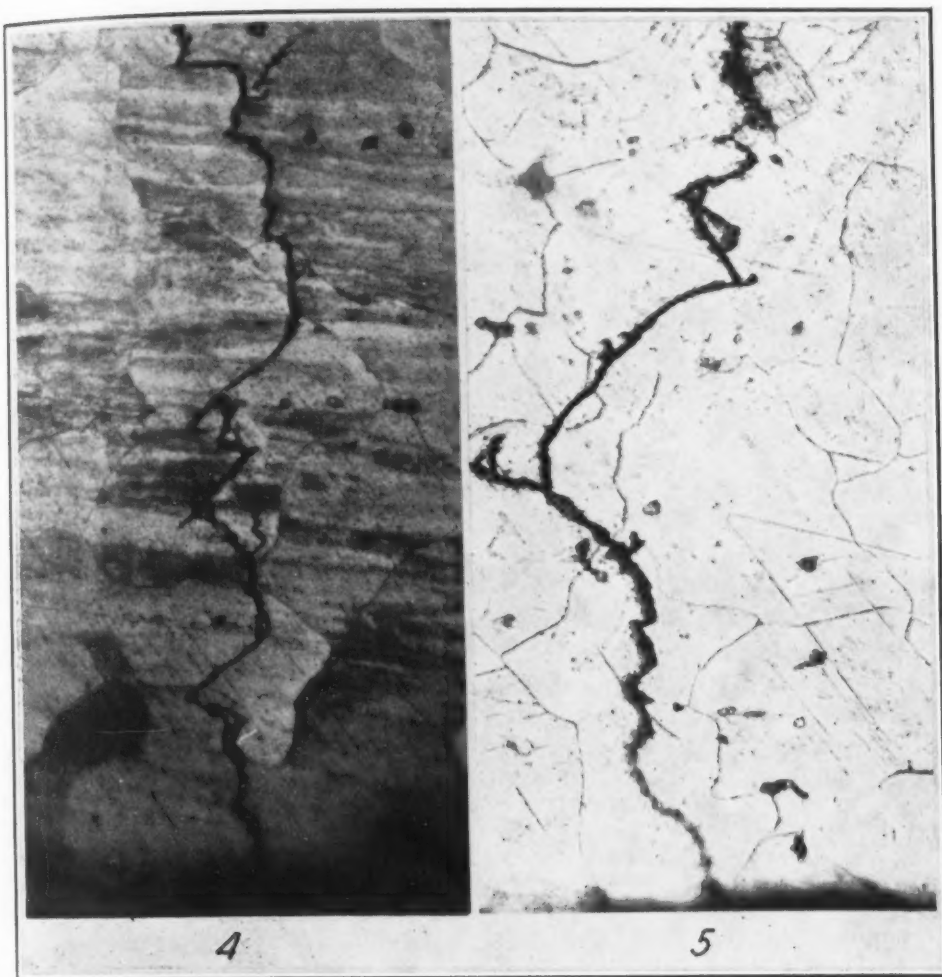


Fig. 4—Fatigue Crack at 0.022-Inch from Original Surface of Armco Iron. Mag. 350x.  
Fig. 5—Fatigue Crack at 0.026-Inch from the Original Surface of Armco Iron. Mag. 350x.

suggests that at these points of change the crack follows the crystallographic planes in the grain and sometimes along the grain boundary. The jagged appearance of the crack and its continuous change of panorama as layers are removed found by the writer in the same materials, fatigued in the same machine as those used by author, are shown in Figs 1 to 6.

From these figures it appears that the influencing factors in the development of a fatigue crack are more complex than stated by the author. The problem is three-dimensional. The writer would suggest that the crystallographic planes in the grain about to crack and the nature of the crack in the preceding grain that failed, as well as the orientation of these grains adjoining and supporting them, together with the weaker bond found between certain dissimilar materials influence the path of the crack.

Figs. 7 to 10 are submitted as a partial answer to a question asked during the discussion of the author's paper as to the nature of fatigue failure in other materials. Fig. 7 shows a fatigue crack in 0.49 per cent carbon steel



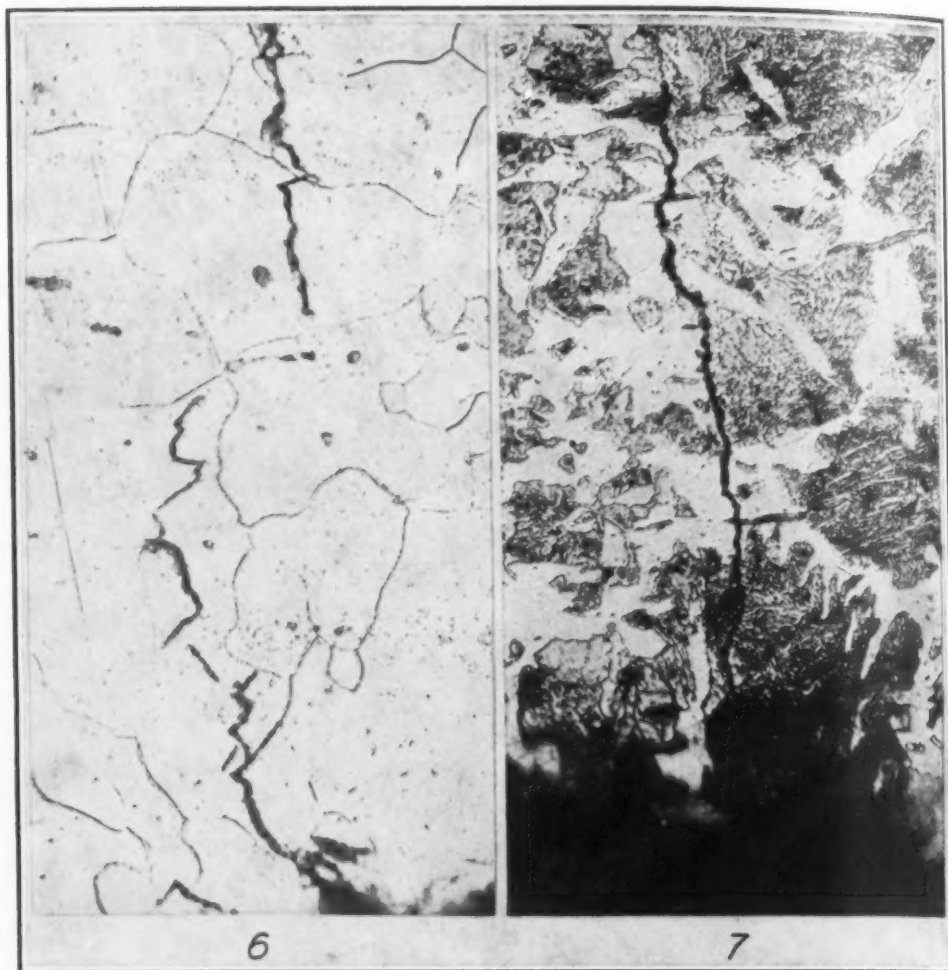


Fig. 6—Fatigue Crack at 0.031-Inch from Original Surface of Armco Iron. Mag. 350x.  
Fig. 7—Fatigue Crack in 0.49 Per Cent Carbon Steel Normalized. Mag. 350x.

normalized, the crack is much straighter than the one found in ARMCO iron. Figs. 8, 9 and 10 (pages 549 and 550) show a fatigue crack in a 0.93 per cent carbon steel normalized. The crack passes around the cementite lamellæ, but seems to favor, in this eutectoid steel, the boundaries between the two constituents.

#### Author's Reply to Discussion

The interest in the preliminary results of this cooperative study, as reflected by the discussion, is gratifying to those who have been interested in carrying out the investigation. The publication of the preliminary results brings forth some confirmatory evidence and also raises several questions.

We are grateful to Dr. Styri for his contribution. He has shown the practical significance of included matter in balls and ball races of bearings and the harmful effects which may be expected. Dr. McAdam, Dr. Styri, and others (who have written us privately), have confirmed the origin of fatigue

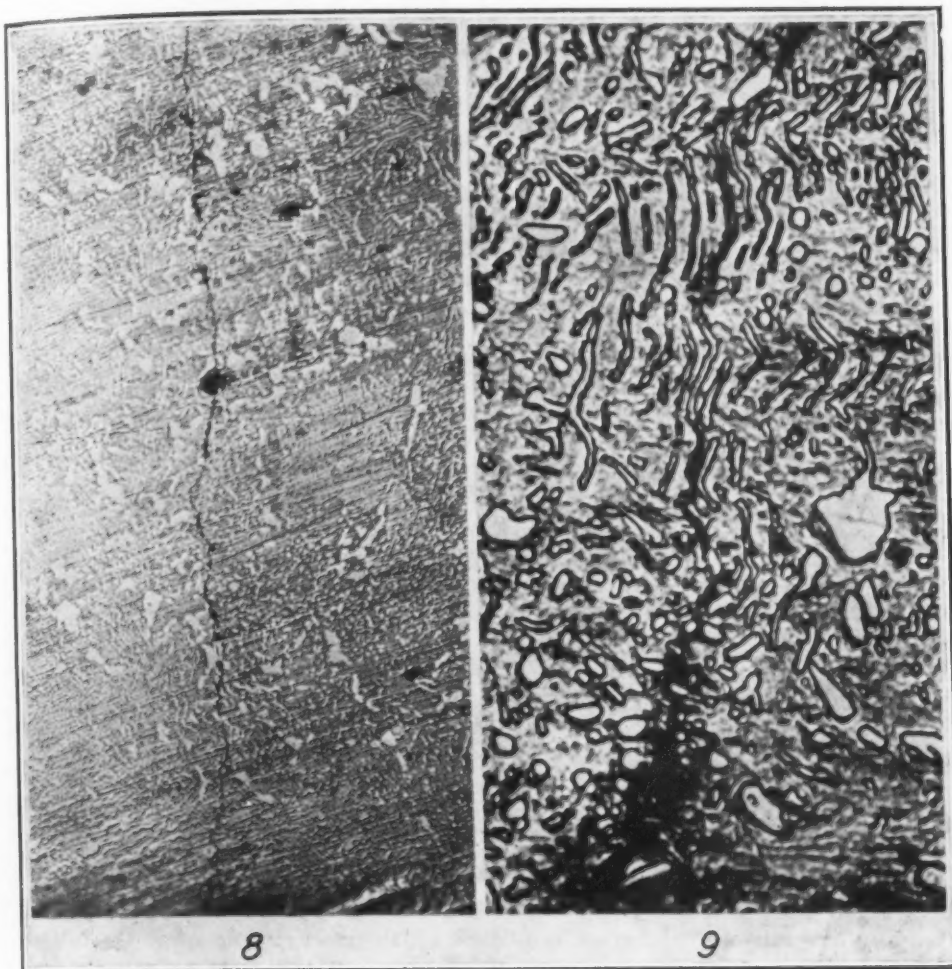


Fig. 8—Fatigue Crack in 0.93 Per Cent Carbon Steel Normalized. Mag. 350x. Fig. 9—Fatigue Crack in 0.93 Per Cent Carbon Steel Normalized. Mag. 2000x. (Muller's Discussion.)

cracks in nonmetallic inclusions. This naturally raises the question as to what happens when nonmetallic inclusions are not present. How do fatigue cracks originate? Are the cracks intercrystalline? Do the grain boundaries then become a potential source of weakness? Can iron be secured sufficiently free from included matter so that the effect arising from this source may be negligible? What will be the effect of grain size and of preferential orientation on the trend of fatigue cracks?

The discussion brings out, we think, the interest which would be manifested in extending the study to include steels which have been heat treated.

The high power microscope; the ultra violet microscope, and the new high power objectives (mono-brom-naphthalene immersion of N. A. 1.60) bring all of these steels, even the finest structures of hardened magnet steels, within our range of vision. The possibilities of the ultra violet microscope and the objectives of N. A. 1.60 have barely been touched upon, but the technique of their application is sufficiently advanced to make them available for ex-



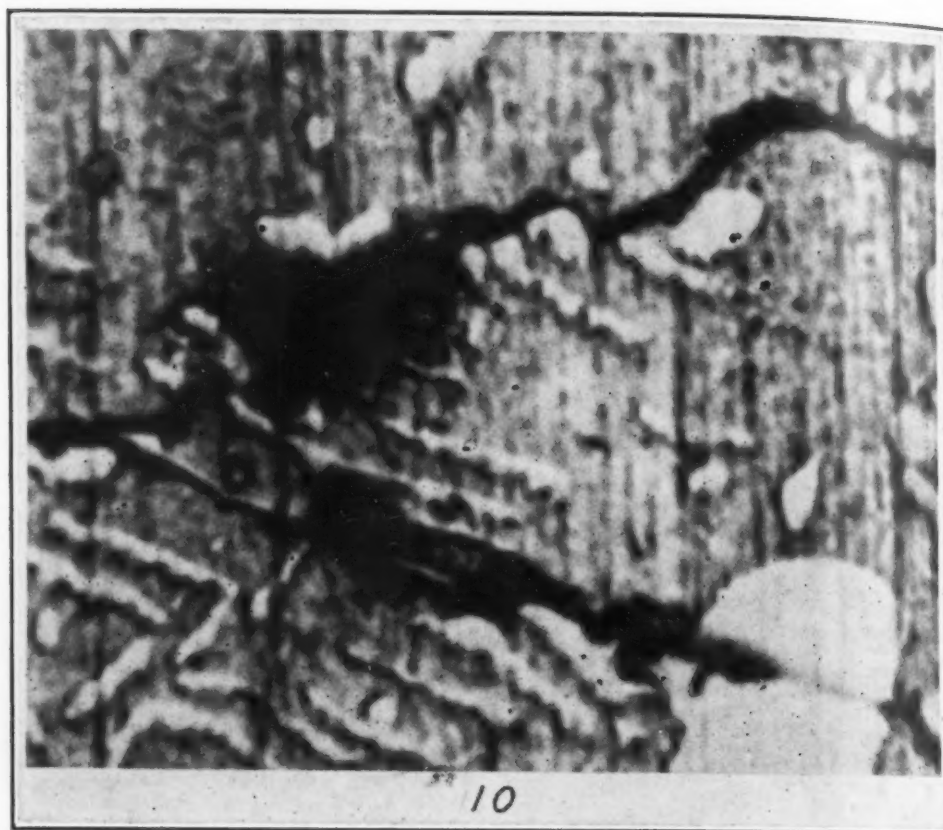


Fig. 10—Fatigue Crack in 0.93 Per Cent Carbon Steel Normalized. Mag. 5000x. (Muller's Discussion.)

tending our range of vision—a condition which was not true when the fatigue work was first undertaken. With the tools available and the technique at least partially developed the limitation now becomes largely one of time available for such investigations.

Mr. J. Muller expresses the view that the high power microscope has placed undue emphasis on the significance of inclusions and that in some cases fatigue cracks ignore some of the inclusions seemingly in the path of the crack.

In the paper it is pointed out that we must think in terms of three dimensions, i. e., conditions above or below the observation plane may have exerted the controlling influence on the trend of the crack. We do think that it would be difficult for one to study the specimens in question and not conclude that nonmetallic inclusions are predisposed to influence the origin and trend of fatigue cracks. In the paper, conclusions are given, supported by evidence, that grain boundaries do not appear to be a potential source of weakness; but if the grain boundaries are conveniently oriented the crack will follow them. There is much to be learned about crystallographic planes. The improved methods at hand for increasing resolution give promise of usefulness in this field.

## SMOOTH FINISH MACHINING OF LOW CARBON PLAIN AND ALLOY STEELS

BY J. S. VANICK AND T. H. WICKENDEN

### Abstract

*Plain and alloy low carbon steels of the carburizing type, have, under incorrect operating conditions, a tendency to machine to a rough finish in final or finishing cuts. It was found that, for each steel and its particular heat treatment, this result was due to the existence of a critical-range of volume removal rates within which a rough finish was obtained. By avoiding this critical range, smoothly finished surfaces could be obtained. Applying the results of this investigation to practice, it was found that cutting conditions leaving a rough surface, could be changed to give a smooth finish by:*

- 1. Either lowering, or preferably increasing the cutting speed until it is outside of the critical range.*
- 2. Maintaining the speed, but changing the cut or feed.*
- 3. Sharpening the cutting angle of the tool and maintaining speed and shape of chip.*
- 4. Changing the hardness of the steel being cut; usually increasing it in order that a good finish is delivered at an easily obtained speed.*

*The following "Introduction" is essentially a synopsis of our work.*

### INTRODUCTION

IT is not infrequently the experience of those who are machining carburizing steels, either of plain carbon or those containing nickel or other alloys, that it is difficult to obtain a smooth machined finish. Under their machining conditions a rough finish is obtained, the tool chip apparently tearing away from the forging rather than shearing off cleanly. It is, however, possible to secure smooth finish machined cuts on such steels and it was the purpose of the investigations described in this article to determine the conditions under which such satisfactory finish cuts can be made.

Ten-inch diameter forged discs of a number of compositions

A paper presented before the Spring Sectional Meeting of the Society at Hartford, Conn., May 20-21, 1926. The authors are associated with the International Nickel Company, New York City.

of plain carbon and alloy carburizing steels were used, the cut being taken on the disc face, the tool starting at the periphery and moving toward the center with continuously decreasing linear speed. In all cases a high speed tool was used having  $\frac{1}{4}$ -inch wide round-nose with a 14-degree side slope, an 8-degree back slope and a 6-degree clearance, these being the angles recommended in Taylor's original work. The cuts taken in this series of tests varied from 0.015 to 0.090 inch in depth and from 0.015 to 0.060 inch in feed. In this manner a good picture could immediately be obtained of the effect of the variation in cutting speeds and dimensions of cut upon the finish obtained. For a fixed feed and cut, the finish on the disc would change from a smooth to rough finish at certain critical speeds, the change being completed within a cutting speed variation of 10 to 15 feet per minute. In general, at a still lower speed the finish would again change from rough to smooth. It was found that this critical area of rough finish on the disc and the range of cutting speeds corresponding to it would vary with the depth of cut, the feed, the hardness of the material, and the cutting angle. Other factors, such as tool dimensions, tool angles, tool supports, coolants, etc., would affect the results but were held constant in these tests.

The rate at which metal is removed in a single tool operation is proportional to the product of feed, depth and linear speed of cut. If we call this "volume removal" rate, the principal conclusion of these tests was that for each steel and heat treatment thereof, using fixed tool dimensions and angles, there is a *critical range of volume-removal rates* within which a rough finish is obtained; increase or decrease of the rate beyond this range, whether by suitable change of feed, depth or speed of cut, will produce again a smooth finish. For example, for a fixed feed and depth of cut, there is a critical speed at which invariably a rough finish will be produced.

A standard nickel carburizing steel SAE-2315, normalized to give a hardness of 166 and with a feed of 0.015 inch and a depth of 0.030 inch, showed a critical range of speed from 60 to 95 lineal feet per minute, within which the finish was invariably rough and unsatisfactory. With the same tool conditions, by either lowering, or preferably increasing the speed, the rough finish could be avoided. A practical indication of the proper speed for a good

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high speed finish was furnished in the appearance of a hot temper-colored chip. The speeds required to obtain a good finish are considerably lower than the breakdown speeds derived from Taylor's work with fine feeds and depths on medium hard steel. Again, for a fixed speed and depth of cut, there is a feed at which a rough finish will be produced. Finally, for a fixed speed and feed there is a depth of cut at which a rough finish will be produced.

The manipulation of feed and depth relation is more difficult to illustrate. In general, if a rough surface were being produced at constant speed, an increase in either feed or depth would increase the chip dimensions and enter a good, "high-speed" finish zone for the larger chip. On the other hand, decreasing the chip dimensions at a constant speed, would correct the trouble by cutting in the "low-speed," "good-finish" zone for the smaller chip. Naturally, all other things being equal, finer feeds left better surfaces than coarse feeds. In addition to the above, the critical range of speed giving a rough finish is apparently lower the higher the Brinell hardness of the steel. Thus, steel No. 2315 mentioned above, which at a Brinell hardness of 166 gave a critical range at 60 to 95 feet per minute and when annealed to give a Brinell hardness of 155, showed a critical range of 100 to 140 feet per minute. The speeds and cuts suitable for obtaining fine finishes are dependent upon the strength and hardness of the individual steels, irrespective of composition. For efficient cutting speeds the limiting hardness is under 220 Brinell and preferably under 200 Brinell. These are hardness values which are readily obtained by normalizing or annealing.

The critical range of cutting speed for rough finish is also reduced by increasing the sharpness of the tool angle. Sharper tool angles such as would be obtained by increasing the cutting angle from 68 to 60 degrees, reduce the high speed necessary for a good finish approximately 30 per cent. Thus S. A. E. steel 2512 having a Brinell hardness of 150 yielded the following typical result:

Cutting Angle	Feed Inches	Cut Inches	Feet per minute					
			Smooth finish at		Poor finish			
			High Speed Min.	High Speed Max.	Low Speed Min.	Low Speed Max.	Speed Min.	Speed Max.
68°	.015	.030	145	285 (x)	0	75	75	145
60°	.015	.030	90	220 (x)	0	45	45	90

(x) Maximum obtained at 108 and 85 revolutions per minute respectively.



Obviously, practical application of the results of these tests may be made in the shop. If it is found in machining carburizing steels that a rough and unsatisfactory finish is being obtained, it is evident that the cutting is being done under a set of conditions which it is possible to correct in one of several ways.

1. Without changing other conditions, by either lowering, or preferably increasing, the cutting speed until it is outside of the critical range.

2. By maintaining the same speed but changing the cut or the feed, thereby obtaining a chip of suitable size.

3. By maintaining the same cutting speed and shape of chip but increasing the cutting angle of the tool and making it sharper.

4. By maintaining the original cutting speed and shape of chip but changing the hardness; usually increasing it in order that a good finish is delivered at an easily obtained speed.

An example of the application of these recommendations, guided by the data listed in a typical table of test results, is included in the text.

#### MATERIALS AND METHODS

The materials chosen for this work consisted of random length billets purchased from mill stock, conforming to the usual Society of Automotive Engineers specifications. No attempt was made

**Table I**  
**Composition, Strength and Hardness of Steels**

No.	S. A. E.	ANALYSIS							
		C	Mn	P	S	Si	Ni	Cr	V
1	1015	.11	.38	.015	.014	.02	....	...	...
2	2315	.19	.57	.040	.015	.25	3.42	...	...
3	2512	.16	.48	.023	.020	.13	4.92	...	...
4	3115	.18	.51	.030	.018	.17	1.48	.68	...
5	5115	.22	.48	.032	.021	.18	....	.91	...
6	6115	.18	.55	.033	.023	.24	....	.92	.21

to make the specifications more rigid in view of the experimental purpose for which the steels were intended and the compositions listed in Table I are indicative of the character of the stock available. Lengths of the 4-inch square billets were upset to form discs, 11 inches in diameter and  $2\frac{1}{2}$  inches thick, which were put through

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**Table II**  
**Strength and Hardness of Steels**

ANNEALED						NORMALIZED					
Lbs. per Sq. In., Per Cent						Lbs. per Sq. In., Per Cent					
No.	S.A.E.	Ten. Str.	Yield Pt.	Elong. in 2"	Red. of Area	B.H.N.	Ten. Str.	Yield Pt.	Elong. in 2"	Red. of Area	B.H.N.
1	1015	49,000 Disc B. H. N.—90;	25,500	41	66	94	52,300 Disc B. H. N.—100;	31,000	41.3	66.6	95
2	2315	78,400 Disc B. H. N.—156;	56,200	33.5	58.3	153	82,700 Disc B. H. N.—166;	61,200	30.5	63.0	158
3	2512	72,200 Disc B. H. N.—149;	54,500	34.3	60.8	140	78,500 Disc B. H. N.—164;	53,900	31.5	61.8	156
4	3115	73,500 Disc B. H. N.—131;	46,800	33.0	59.9	138	73,600 Disc B. H. N.—148;	48,900	31.8	69.0	143
5	5115	68,400 Disc B. H. N.—130;	37,000	31.8	53.3	131	71,600 Disc B. H. N.—148;	43,300	32.0	67.5	140
6	6115	68,300 Disc B. H. N.—132;	41,200	33.0	64.8	131	73,600 Disc B. H. N.—152;	49,400	34.7	69.5	143
Heated 2 hours at 1650° F.; furnace cooled.						Heated 2 hours at 1650° F.; air cooled.					

**Table III**  
**Strength and Hardness of Steels**

QUENCHED and TEMPERED						QUENCHED					
Lbs. per Sq. In., Per Cent						Lbs. per Sq. In., Per Cent					
No.	S.A.E.	Ten. Str.	Yield Pt.	Elong. in 2"	Red. of Area	B.H.N.	Ten. Str.	Yield Pt.	Elong. in 2"	Red. of Area	B.H.N.
1	1015	55,000 Disc B. H. N.—106;	36,900	36.5	71.0	110	60,800 Disc B. H. N.—127;	38,200	32	67.7	121
2	2315	90,300 Disc B. H. N.—195;	65,500	29.5	71.4	183	..... Disc B. H. N.—321;	.....	.....	.....	.....
3	2512	87,700 Disc B. H. N.—190;	64,900	29.8	70.2	179	75,900 ✓ Heat treated Disc B. H. N.—166	53,700	34.0	63.9	156
4	3115	78,400 Disc B. H. N.—178;	56,700	27.5	70.9	170	.....	.....	.....	.....	.....
5	5115	82,650 Disc B. H. N.—183;	60,450	26.0	72.7	170	.....	.....	.....	.....	.....
6	6115	113,700 Disc B. H. N.—271;	95,000	21.0	59.6	241	.....	.....	.....	.....	.....
Heated 2 hours at 1650° F.; water quenched. Reheated to 1200° F.; 2 hours.						Quenched in water from 1650° F. (xx) Heated to 1650° F., furnace cooled to 800° F., cooled in air. ✓ Quenched Disc B. H. N. 364; Shore 59.					

several heat treatments to represent typical conditions of the steel upon its entry to the machining operation.

Since each successive test depended upon obtaining the optimum speed for a given feed and depth of cut, the circular disc method of testing was adopted. By machining the face of the disc running at a constant speed the cutting speed is automatically varied from maximum to minimum as the tool moves from edge



to center. If a smooth surface is produced at some position on the face of the disc its position can be measured, and knowing the speed the corresponding speed may be determined.

A 10-inch diameter disc which would allow a peripheral speed of ten times the speed of a 1-inch disc was used, i. e., at 30 feet per minute for the 1-inch diameter the 10-inch diameter circle would be moving at 300 feet per minute. A graphical diagram

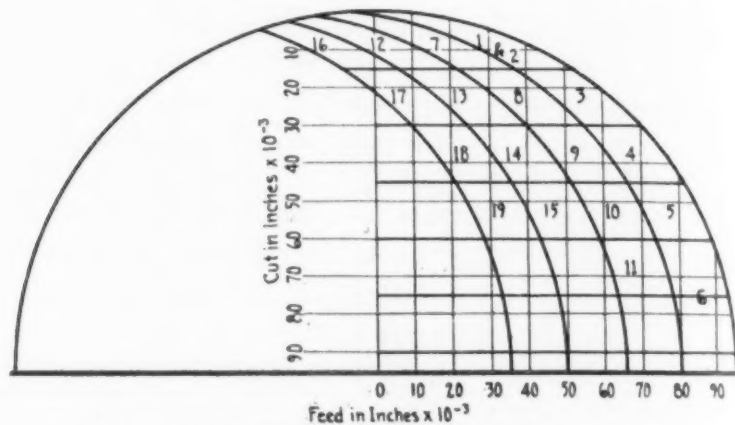


Fig. 1—Contour of Cutting Edge and Shape of Chip.

of the change in cutting speed for the different speeds of the lathes was used to determine the speed of cutting for any of the gear-shifts in the two lathes, as well as the speed position on the surface of the discs, of changes in "finish."

Four discs of each steel were made for the test to represent annealed, normalized, quenched, and quench-tempered conditions. The strength and hardness of each disc was determined for the several heat treatments as a part of the cutting program. The condensed results are listed in Tables II and III. The lathes used for these tests consisted of 24-inch and 15-inch Le Blonde machines, both of the heavy-duty type. These lathes were motor driven at a constant speed and geared to a broad range of spindle speeds.

The cutting tools used for these tests consisted of  $\frac{1}{2} \times \frac{1}{2} \times 6$ -inch lengths of high speed steel, forged to  $\frac{1}{4}$ -inch wide and ground to a  $\frac{3}{16}$ -inch diameter nose of the contour shown in Fig. 1. The tool is essentially one of the Taylor type,<sup>1</sup> shaped to the angles which Taylor had recommended. This shape is identical with the form of the round-nosed, roughing tool and consists of a 68-degree cutting angle with an 8-degree back slope, a 14-degree side slope,

<sup>1</sup>Taylor, F. W. The Art of Cutting Metals; Transactions, American Society of Mechanical Engineers. Vol. 28, 1907, p. 31.

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Cut Num- ber	Depth
1	.015
2	.015
3	.030
4	.045
5	.060
6	.090
7	.015
8	.030
9	.045
10	.060
11	.090
12	.015
13	.030
14	.045
15	.060
16	.015
17	.030
18	.045
19	.060

1	.015
2	.015
3	.030
4	.045
5	.060
6	.090
7	.015
8	.030
9	.045
10	.060
11	.090
12	.015
13	.030
14	.045
15	.060
16	.015
17	.030
18	.045
19	.060

<sup>2</sup>Kent.

and a 6-degree end clearance angles. A special tool holder was made which fitted snugly in the tool post and clamped close to the cross feed carriage to reduce vibration.

A preliminary test on some 0.35 per cent carbon steel showed that a good finish could be obtained upon this material if light feeds and high speeds were used. The range of speed was quite broad, but narrowed rapidly as the feed and depth of cut were increased. These preliminary tests showed that feed, speed and cut were inter-related and the best plan to follow would be one that included the working range for these variables.

The working range for high speed lathe tools is listed in the handbooks.<sup>2</sup> The speed-feed-depth relations for soft materials,

Table IV  
Cutting Tests on S. A. E. 3115

ANALYSIS							BHN = 130. FURNACE ANNEALED FROM 1600° F.			
C	Mn	P	S	Si	Ni	Cr				
.18	.51	.030	.018	.17	1.48	.68				
Cut Num- ber	Depth	Feed	R. P. M.	Colored Chip Speed	Smooth Finish at High Speed ft. per minute		Smooth Finish at Low Speed ft. per minute		Speed Producing Poor Finish ft. per minute	
					Min.	Max.	Min.	Max.	Min.	Max.
1	.015	.015	10	...	...	...	0	20	20	25
2	.015	.015	152	...	160	400	0	120	120	160
3	.030	.015	108	105	145	285	0	75	75	145
4	.045	.015	85	110	110	220	0	65	65	110
5	.060	.015	60	75	70	155	0	40	40	70
6	.090	.015	77	85	95	205	..	..	0	95
7	.015	.030	152	160	160	400	0	100	100	160
8	.030	.030	108	..	115	285	0	45	45	115
9	.045	.030	41	65	60	110	0	25	25	60
10	.060	.030	29	50	45	75	0	25	25	45
11	.090	.030	77	85	65	205	..	..	0	65
12	.015	.045	152	160	140	400	0	85	85	140
13	.030	.045	108	115	115	285	0	45	45	115
14	.045	.045	77	85	85	205	0	45	45	85
15	.060	.045	29	..	55	75	0	55	..	..
16	.015	.060	152	120	120	400	..	..	0	120
17	.030	.060	108	85	115	285	..	..	0	115
18	.045	.060	77	65	95	205	..	..	0	95
19	.060	.060	54	100	100	140	..	..	0	100

<sup>2</sup>Kent. P. 1266—9th Edition. Machinery. P. 796—5th Edition.

Table IV  
Cutting Tests on S. A. E. 3115—Continued

BHN = 150. NORMALIZED FROM 1600° F.										
Cut Number	Depth	Feed	R. P. M.	Colored Chip Speed	Smooth Finish at High Speed ft. per minute		Smooth Finish at Low Speed ft. per minute		Speed Producing Poor Finish ft. per minute	
					Min.	Max.	Min.	Max.	Min.	Max.
1	.015	.015	20	..	..	..	..	..	0	55
			54	..	..	..	60	90	0	140
2	.015	.015	152	180	160	400	0	90	90	160
3	.030	.015	108	145	140	285	0	45	45	140
4	.045	.015	60	110	95	155	0	65	65	95
5	.060	.015	60	95	85	155	0	40	40	85
6	.090	.015	77	85	95	205	0	45	45	95
7	.015	.030	152	105	120	400	0	55	55	120
8	.030	.030	108	85	85	285	0	70	70	85
9	.045	.030	41	65	65	110	0	25	25	65
10	.060	.030	29	60	55	75	0	30	30	55
11	.090	.030	77	45	55	205	..	..	0	55
12	.015	.045	152	120	160	400	120	160	0	120
13	.030	.045	108	100	100	285	0	45	45	100
14	.045	.045	77	65	65	205	..	..	0	65
15	.060	.045								
16	.015	.060	77	85	85	205	..	..	0	85
17	.030	.060	108	60	100	285	..	..	0	100
			77	55	95	205	..	..	0	95
18	.045	.060	54	50	75	145	50	75	0	50
19	.060	.060	29	40	45	75	..	..	0	45

taken from Taylor's work and listed in Kent's Handbook, were chosen. The size of tool used was approximately one-half of that of the smallest tool listed in the tables and the dimensional relations required that about 10 per cent be discounted from the listed speeds. In addition to the three classifications given in the handbooks an interpolation had been made between the medium-soft steels and the medium-hard steels, giving two additional tables of speeds for the different sizes of cut. The series of approximately 20 cuts which were taken from each disc are listed in Table IV, which is typical of the data obtained. Cut No. 16—corrected for change in tool size—was used as a speed index, since the dimensions of the chip corresponded to handbook speeds for the 1/2-inch tool when cutting soft, medium, and hard steels. In the large

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Table IV  
Cutting Tests on S. A. E. 3115—Continued

BHN = 180. QUENCHED FROM 1600° F.; TEMPERED AT 1200° F.										
Cut Num- ber	Depth	Feed	R. P. M.	Colored Chip Speed	Smooth Finish at High Speed		Smooth Finish at Low Speed		Speed Producing Poor Finish	
					ft. per minute Min.	Max.	ft. per minute Min.	Max.	ft. per minute Min.	Max.
1	.015	.015	29	..	..	..	0	40	40	75
2	.015	.015	85	95	160	225	0	70	70	160
3	.030	.015	60	80	110	160	40	110	0	40
4	.045	.015	60	80	55	155	0	0	0	55
5	.060	.015	60	70	40	155	..	..	0	40
6	.090	.015	41	70	60	110	0	25	25	60
7	.015	.030	85	70	95	225	..	..	0	95
8	.030	.030	60	50	75	155	..	..	0	75
9	.045	.030	41	45	60	110	30	60	0	30
10	.060	.030	29	35	55	75	0	20	20	55
11	.090	.030	41	45	45	110	..	..	0	45
12	.015	.045	60	60	60	155	..	..	0	60
13	.030	.045	41	35	65	110	..	..	0	65
14	.045	.045	41	40	55	110	..	..	0	55
15	.060	.045	29	30	40	75	..	..	0	40
16	.015	.060	60	75	80	155	..	..	0	80
17	.030	.060	60	30	60	155	..	..	0	60
18	.045	.060	41	35	..	..	..	..	0	110
19	.060	.060	29	25	40	75	..	..	0	40

lathe, if excessive speeds upon too-heavy cuts were used, the tool would break down, while in the light lathe the same factors would stall the motor. The speed values given in the handbooks were used as maxima except in such cases where no values were given, when the square-root law,

$$S = \frac{1}{\sqrt{F}}$$

was used. The values derived were expected, as stated in Kent, to allow a tool endurance of 11½ hours. For most cuts the cutting speed was less than 300 feet per minute and a single cut across the face could be taken. When the maximum allowable cutting speed exceeded 300 feet per minute, two cuts at a low and high speed were taken, to cover the range. The practice of following a coarse cut with a fine one, viz.: No. 9, followed by



No. 6 or No. 2, did not yield significantly different results than if cuts No. 6 or No. 2 were put through, irrespective of previous finish. In general, the cuts were taken in the serial order in which they are listed in the test data.

#### EXPERIMENTAL RESULTS

All of the steels tested showed critical ranges of cutting speed which gave rough surfaces. The speed ranges were closely related to the strength and hardness of the piece, irrespective of composition. Since the data possess the above general relation and the results are applicable to a  $\frac{1}{4}$ -inch tool, shaped as described, only Table IV is presented as typical of the values obtained. The translation of the data which have been obtained into current shop practice where changes in tool size, tool angles, contour of edge, tool support, limiting speeds, power available, and other factors come into play, would probably require more work than a repetition of this test guided by the results of a sample table of data. For this reason, more than five tables of data similar to that of Table IV have been omitted.

Probably the most important conclusion reached as a result of these tests was that for any particular steel and heat treatment thereof, together with fixed feed and depth of cut, there is a critical speed at which invariably a rough finish will be produced. At speeds greater and less than this critical range of speed the finish produced is smooth and satisfactory. The actual value of this critical range of speed varies, of course, with different steels and with the different machining conditions, particularly depth of cut and feed, shape of tool, etc. The transition from rough to smooth finish is accomplished by a change in the cutting speed of as little as 10 to 15 feet per minute and is quite sharp and definite. The high speeds for any given size and shape of cut produce the best finish. A high speed surface is smooth, relatively clear of slivers or checks and free from torn or gouged spots. An examination of Table IV shows that the best high speeds for a smooth finish occur from 40 feet per minute for the coarse cuts to over 100 feet per minute for the fine cuts. Some explanation for this is found in the relations between the volume of shaving and the contour of the cutting tool engaged in the work. An inspection of Fig. 1 will show that the cuts of the size mentioned engage a comparatively large portion of the edge, therefore the work of removing

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metal is distributed over a long edge. As the depth of cut is increased to engage zones 4, 5 and 6 of Fig. 1, the portion of cutting edge which is engaged in the work is appreciably lessened and the ratio of metal area to length of cutting edge appreciably

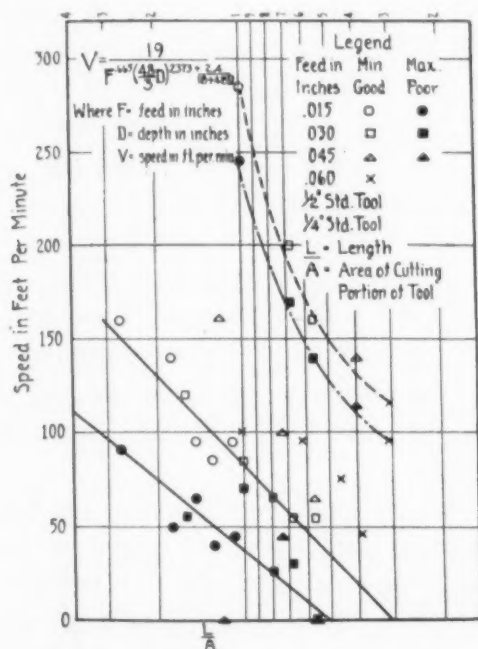


Fig. 2—Diagram of Speed Versus Chip Sizes Blocking Out "Poor-Finish" Speeds Between the Heavy, Straight Lines. Lower Straight Line Represents Maximum, Low-speed, for Good Finish; Upper Straight Line Represents Minimum High Speed, for Good Finish; Broken Lines Above are Computed from Taylor's Data and Show a Wide Margin Between Breakdown Speed and Good Finish Speeds.

reduced. For a feed of 0.015-inch the thickness of the shaving represented by the areas 1 to 6 of Fig. 2, is comparatively small and the "feed-movement" of the tool is such as to put the work of finishing on the cutting edge bounding areas 1 to 3.

Somewhat similar relations are developed when the feed is increased. As the feed is increased from 0.015 to 0.060-inch it may be readily seen from an examination of Fig. 1 that a large portion of the edge of the tool is engaged in cutting as long as the depth of cut remains shallow. An inspection of Fig. 1 shows that the chip area for the coarser feeds grows rapidly as the cut is deepened, while the length of the edge engaged increases slowly. It is obvious that a larger feed for the shape of the tool used will leave an increasingly broader trough behind tool. Where this



trough-like surface can be accepted, a smooth finish may be obtained at comparatively high speeds. An examination of Table IV shows again that these speeds generally range from 50 to 150 feet per minute, depending upon the dimensions of the cut for the annealed and normalized steels. Thus, for a standard nickel carburizing steel SAE-2315, heat treated to give a hardness of 195 and with a feed of 0.015-inch and a depth of 0.030-inch the critical range of speed within which the finish was invariably rough and unsatisfactory was from 40 to 105 lineal feet per minute. As shown in Table IV, four different feeds were used against a series of progressively deeper cuts, the whole forming an interlocking series which might be arranged with either feed or depth as one constant factor while the other increased in sequence. When the results for any steel are plotted, the points assume the form shown in Fig. 2. It may be observed that the smooth-finish speed decreases as the chip size increases, whether the finish be the high or the low-speed type.

The speeds at which good surfaces are obtained on the soft steels appear excessively high when compared to the roughing cut speeds. Taylor made the statement that no increase in speed was obtained on hard steels or irons when cuts finer than those of his "roughing cut test" were used. He did show that very high speeds are possible on soft materials. As an illustration, his work on break-down speeds versus depth and cut are calculated from his data and formula shown in Fig. 2. The original curve represents values obtained with a  $\frac{1}{2}$ -inch tool shaped to his standard angles taking 0.083-inch cut at increasing feeds from a "medium-hard" steel of the following properties:—

Composition		Physical Properties	
C	0.34	Tensile	70,290 lbs. per sq. in.
Mn	0.54	Elastic limit	34,600 lbs. per sq. in.
P	0.037	Elongation	29.0 per cent
S	0.026	Reduction of area	44.0 per cent
Si	0.17	BHN	130 (estimated)

The derived values shown in the lower curve represent a 15 per cent discount in the speed which is necessary to balance the change in tool dimensions, from a  $\frac{1}{2}$ -inch tool to a  $\frac{1}{4}$ -inch tool. The position of the curves of break-down speed leave a comfortable margin between them and the high speeds for good "finish."

Early in the work, the observation was made that the termination of the smooth finish zone at the high speed or outer por-

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tion of the disc was accompanied by the disappearance of "temper-color" in the removed chip. An attempt was consequently made to record the position of this effect. The rough agreement of the speed position of the "temper-colored" chip with that of the minimum, high speed for good finish, affords a practical index of a proper speed for good finish. This rule is especially true of light feeds and depths of cut.

All cutting operations are liable to chatter because of vibrations produced in the machine tool. These vibrations are conveyed to the junction of the tool and the work and recorded on the cut surface. An overloaded tool will give rise to a series of vibrations depending upon the elasticity and "play" in the tool, holder, carriage, lathe bed, gears and spindles. The effect is distinctly different from the roughness produced at incorrect speeds. For example, the coarse feed cuts upon the quenched steels, showed chatter-marks in the glazed, brilliantly smooth finish. An artificial production of chatter was undertaken by making the distance from support to tool apex 3 inches instead of 1 inch, and the series of cuts begun on a normalized S. A. E. 2315 steel. With this increase in the lever arm, the tool immediately showed signs of distress by setting up a persistent howl which apparently conformed to the frequency of the vibrations. Only the light cuts (2, 3 and 8) were completed. In taking cut No. 8, the tool rose approximately 50 per cent out of the depth of cut and rode over the intermediate (middle  $\frac{1}{3}$ ) of its traverse, retaining, at the same time, the "speed-finish" characteristics corresponding to those yielded by a non-chattering tool. In attempting to take cut No. 9 the lathe was shaken badly and the tool was broken down at a speed of 205 feet per minute, which was considerably slower than the speed upon a successful cut obtained with a short length tool. Obviously, heavy finishing cuts would have ruined the tool as well as the work and none were attempted. These few cuts clearly indicated that chatter recorded its effects independently and that excessive chatter or lathe vibration during the course of cutting would be instantly recognized by the howl of the tool, accompanied by rapid, tool breakdown.

In taking finishing cuts at large feeds the appearance of the surface left by the advancing tool becomes a function of the vertical profile of the cutting edge. It may readily be seen, from

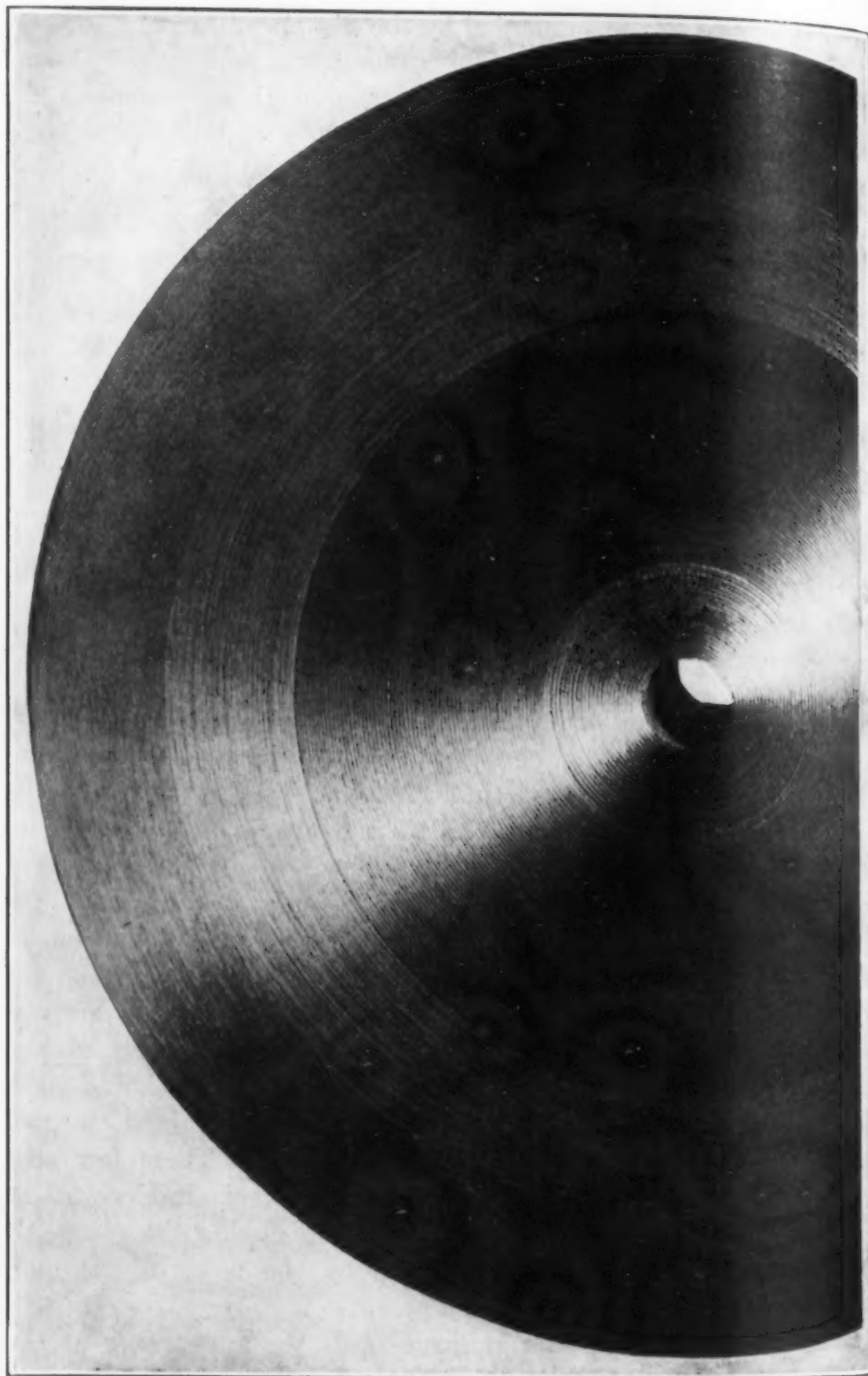


Fig 3—Illustration of Smooth and Rough Zones Produced at Will on a 5-inch Radius Disc of Steel S. A. E. 3115 Annealed—Outer smooth (4 to 5 inch from center) zone, represents 0.030 inch depth x 0.015 inch feed at 180 to 225 ft. per min. Outer rough (3 to 4 inch from center) zone, represents 0.030 inch depth x 0.015 inch feed at 95 to 125 ft. per min. Inner Smooth (1 to 3 inch from center) zone, represents 0.030 inch depth x 0.030 inch feed at 90 to 270 ft. per min. Inner rough (0 to 1 inch from center) zone, represents 0.030 inch depth x 0.030 inch feed at 0 to 25 ft. per min.

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an inspection of Fig. 1, that any feed which advances the tool appreciably beyond the apex of the arc at the nose, leaves an increasingly wider groove or trough in the wake of the cut. Since a finer feed would correct this grooved appearance and produce a smoother finish, one of the manipulations in feed and cut relations could be applied and an equal volume of metal removed with a lighter feed and deeper cut. The ridges forming the walls of the grooves can also be reduced or eliminated by prolonging the horizontal projection of the trailing side of the tool, enough to provide for the maximum feed, thus cleaning up the residue left by the curved edge. This shape is in general usage for finishing tools. The "finish" in the wake of the tool, in this case, remains closely related to the speed, cutting angle, depth and size of cut taken by the rounded side of the tool. The advantages of the round nose tool for roughing cuts has been amply discussed in Taylor's work. Briefly, it distributes the work of cutting, thereby reducing the danger of breakdown at a corner or at some overloaded point on the edge. In finishing tools, the same principle, directed toward relieving the work on the apex of the nose should permit higher finishing speeds at that point with their characteristically smoother surface. This is accomplished by maintaining a fine feed which engages as little of the zone 1 to 3 area of Fig. 1, and as much of the zone 4 to 6 area as possible. The apex edge of the tool, engaged in "finishing" would remain long in proportion to the area of chip it was removing and the trace in the wake of the tool would be narrow and more pleasing in appearance. For these reasons, it is apparent that a deep cut with a fine feed and high speed is preferable to a shallow cut, with a coarse feed and lower speed, where the volume of metal to be removed is the same.

Another very interesting and important conclusion shows that the critical range of speed giving a rough finish is apparently lower, the higher the Brinell hardness of the steel. The cutting data in Table IV clearly shows that "good finish" surfaces may be obtained at lower speeds as the hardness of the steel increases. Thus, steel S. A. E. 2315, mentioned above, which at a Brinell hardness of 195 gave a critical range at 40 to 105 feet per minute, when annealed to give a Brinell hardness of 155, showed a critical range of 100 to 140 feet per minute. This relation to finish has an immensely practical value. For example, where the speed of



the tool cannot be increased for practical reasons, it is evident that the work may be slightly hardened. In the case of steel S. A. E. 2512, the normalized steel and the same steel treated as directed in A. S. S. T. "recommended practice for camshafts" possessed identical properties and machined equally well in the cutting tests. It is quite possible that the larger section of the pieces treated in these tests balanced the softening which "furnace cooling to 800 degrees Fahr." would produce in camshafts. The range of hardness explored in this work extended from the 90 Brinell hardness number of the annealed low carbon steel to 360 Brinell hardness number in some of the quenched steels. The cutting speeds for good finish over this range appeared to be dependent upon the strength and hardness and independent of the composition. The strength of these steels increases with their hardness and either factor may be used to control the cutting speed. Some tendency appears for the tensile strength to be the dominant factor in light cuts upon steels of equal hardness but the difference is small and the effect disappears in heavier cuts.

The extremely high speeds, such as 780 feet per minute, which could be successfully applied to annealed, very soft (90 Brinell hardness number), plain carbon steel, illustrate the broad range of speed available above that required for smooth cutting. As the steel became harder, the range between good finish, high speed and the speed at which the tool edge was destroyed became narrower. Good surfaces were easily obtained in its quenched, 130 Brinell hardness number form at practical cutting speeds. The alloy steels in their quenched condition attained a hardness exceeding 300 Brinell hardness number, a value at which practical machining operations on this type of steel would not be undertaken. Tests upon a few of the quenched discs representing a hardness in excess of 320 Brinell hardness number were limited to a few light cuts which produced a glazed finish and showed that the cutting required too much from the lathe and tool. For example, the hardness of 250 to 270 Brinell hardness number obtaining in the heat treated S. A. E. 6115 steel, compared to the 150 Brinell hardness number obtaining in the same steel, normalized, slowed the smooth finish cutting speed from a range of 110 to 225 feet per minute to a range of 25 to 50 feet per minute in the heavier cuts. For efficient cutting, the limiting hardness is under 220 Brinell

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hardness number, and preferably under 200 Brinell hardness number. These are hardness values which are readily obtained by normalizing or annealing.

The critical range of cutting speed for rough finish is also reduced by increasing the sharpness of the tool angle. Thus, increasing the cutting angle from 68 to 60 degrees reduced the speed necessary to secure a good finish by at least 30 per cent in the case of S. A. E. steel 2512 having a Brinell hardness number of 150. The side slope of the edge had been increased from the 14 degrees recommended for hard materials to 22 degrees recommended for soft materials. In addition to the change in speed relations, the chip was coiled in a spiral of small diameter and left the work at an acute angle to the line of cutting. It is evident that sharpening the cutting angle offers a means of lowering the minimum speed for "good finish" and of controlling the shape and direction of discharge of the removed chip.

Obviously, practical application of the results of these tests may be made in the shop. If it is found in machining carburizing steels that a rough and unsatisfactory finish is being obtained, it is evident that the cutting is being done under one of the unfavorable conditions referred to above and it is possible to correct it in one of several ways.

1. Without changing other conditions, by either lowering, or preferably increasing, the cutting speed until it is outside of the critical range.

2. By maintaining the same speed but changing the cut.

3. By maintaining the original cutting speed and shape of chip but changing the hardness; usually increasing it in order that a good finish is obtained at an easily obtained speed.

4. By maintaining the same cutting speed and shape of chip but increasing the cutting angle of the tool and making it sharper.

The practical application of the data of Table IV may be illustrated in the following example. Suppose a finishing cut of the dimensions of cut No. 7 (Table IV) is recommended upon a piece limited by other factors (such as equipment, or size of piece, etc.) to a surface speed of 125 feet per minute. Then cut No. 7, taken from the annealed steel, would produce a rough, torn, surface since its poor finish speed is embraced in the 100-160 feet per minute range. Assuming that the speed cannot be increased above



125 feet per minute, it would be possible, in a first attempt, to avoid the rough-finish speed by dropping to 100 feet per minute or less and operating in the smooth-finish, low-speed range at some sacrifice of production.

To avoid reducing production, a change in the feed and cut relation upon the same steel could be applied as a second expedient, by using cuts No. 8 or No. 4 at the high speed, in finishing the cutting schedule. Assuming that the finishing cut is limited in depth, then coarser feeds may be applied until a "smooth finish" feed is obtained for the limited speed: viz., the use of cut No. 9 in place of cut No. 4 above. This manipulation would increase production. If available, higher speeds, within the endurance of the tool, might be used as well.

In a third effort, the initial speed and cut (No. 7) could be retained but the material hardened to 180 Brinell hardness number which as the tests upon the heat treated piece show, should finish smooth.

Finally, the sharpening of the cutting angle with its attendant reduction of the critical speeds could be utilized in taking the original (No. 7) cut upon the annealed steel resulting in lowering the poor finish zone under 110 feet per minute and producing a smooth finish above this speed.

In conclusion, the helpful advice and criticism of Dr. P. D. Merica in his constant contact with this work, is gratefully acknowledged.

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## BASIC OPEN-HEARTH PRACTICE

By C. H. HERTY, JR.

### *Abstract*

*Open-hearth practice for three types of charges are compared, three heats of steel being used for the comparison. The first two are pig iron and scrap heats with no ore in the charge, the first of these charged with very large scrap, the second with medium scrap. The third heat is an iron ore heat, 20,000 pounds of ore being charged with the scrap.*

*The comparison covers the melting period, the lime boil, the working period and the efficiency of the de-oxidizers used. The effect of sulphur in the gas on the sulphur content of the metal is discussed and illustrated.*

THE basic open-hearth process for the refining of pig iron consists, with the exception of the removal of sulphur, of an oxidation of the impurities, and some of the iron, followed by the removal of the oxide formed either by combination or solution in the basic slag or by the removal of the oxide as a gas.

Silicon and phosphorus in the iron are oxidized and their oxides combine with the basic components in the slag. Manganese and iron are oxidized and their oxides dissolve in the slag, though it is probable that a part of these oxides exist in the slag as silicates. Carbon is oxidized to carbon monoxide,  $\text{CO}$ , which escapes from the bath, burning above the slag to  $\text{CO}_2$ . The element sulphur is removed by reaction with the lime in the slag, forming  $\text{CaS}$  and  $\text{CaSO}_4$ , and the removal is retarded by the oxidation of the slag. Thus four of the five elements to be removed from the iron depend on the oxidation of the slag for their removal while the removal of the fifth is held back by this condition. It is therefore necessary to strike a balance on degree of oxidation for speed of working and extent of elimination of the metalloids, particularly carbon, phosphorus and sulphur.

Furthermore, since the iron oxide is soluble in the metallic bath, the four oxidizing reactions are much faster than the re-

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moval of sulphur, since these reactions are carried out in both a single phase and at the slag-metal surface, whereas the elimination of sulphur takes place at the slag-metal surface only.

The basic slag on which the process depends, is obtained by charging limestone or lime, the lime from either source combining with the acid oxides to form the slag. The removal of sulphur and phosphorus depends, in addition to the oxidation of the slag, on the basicity, which varies with the amount and composition of the metal, with the amount of lime or limestone charged and with the erosion of the banks of the furnace.

Before taking up working of the three tests described in this paper, it will be well to briefly show the function of each impurity in the metal and of the most important constituents of the slag.

#### METALLOIDS IN THE PIG IRON

##### *Carbon*

Carbon is usually the controlling factor in determining the speed at which a given heat of steel is made. Compared to silicon, manganese and phosphorus, its oxidation is slow and in order to eliminate it rapidly it is necessary to carry a very fluid and highly oxidized slag, or to control the charge so that the iron oxide formed during melting and during the lime boil is just sufficient to eliminate all the metalloids, including carbon, to the point desired without "working" the heat.

##### *Silicon*

The silica resulting from the oxidation of the silicon in the iron combines with lime in the charge to form the basis of the slag. If the silicon in the charge is too low the slag will be "heavy" and slow working, whereas if the silicon is too high the slag will be too acid and will require excessive amounts of lime, the erosion of the furnace banks will be increased, and the removal of phosphorus and sulphur will be greatly retarded. The best range of silicon in pig iron is from 0.85 per cent to 1.35 per cent.

##### *Manganese*

Manganese in pig iron serves three useful purposes. First, with high percentages of manganese, 1.50 to 2.00 per cent, a

considerable amount of sulphur will be eliminated from the iron between the time it leaves the blast furnace and the time it is introduced into the open-hearth furnace. Second, the fluidity of the slag in the open-hearth furnace is increased with increasing manganese content and this makes for faster working of the furnace. Third, this high "residual manganese", i. e., manganese content of the steel before the final additions are made, gives a pronounced saving in ferromanganese. Opposed to these advantages must be placed the fact that with high manganese iron, the amount of manganese oxide,  $MnO$ , dissolved in the steel will be higher than with low manganese iron and this may in some cases cause imperfections in the finished steel. The present tendency in steel making is to use a high manganese iron.

#### *Phosphorus*

If the iron is not too high in phosphorus, say 0.300 per cent, little difficulty is experienced in eliminating this element provided the slag is kept sufficiently basic and oxidizing. With high phosphorus iron, 0.80 to 1.00 per cent, the straight stationary process cannot be used, and a process in which the slag may be removed when necessary must take its place.

#### *Sulphur*

This element is the most difficult of the five metalloids to remove. This is due first to the fact that the elimination is slow, the reaction taking place at the slag-metal surface only, and second that the oxidizing slag retards the reaction, and third to the absorption of sulphur from the furnace gases, if the gases go higher than a certain sulphur content, this content depending on the sulphur content of the scrap and of the slag. With producer gas and coke-oven gas the scrap and slag will almost always absorb the sulphur, with tar and oil containing 0.50 per cent sulphur or under the reverse is usually the case, i. e., the scrap and slag lose sulphur to the gas, while with natural gas, sulphur will always pass from the scrap or slag into the gas.

#### SLAG CONSTITUENTS

##### *Iron Oxides ( $FeO$ and $Fe_2O_3$ )*

As has been shown above iron oxide is the primary agent



in the removal of four of the five metalloids in pig iron. Iron oxide in the slag exists in two forms, ferrous oxide,  $\text{FeO}$ , and ferric oxide,  $\text{Fe}_2\text{O}_3$ . The ferric oxide is the result of oxidation of  $\text{FeO}$  by the furnace gases.

Iron oxide in the slag is derived from four sources,

1. Oxidation of the scrap during melting.
2. Oxidation of the bath by the gases from the decomposition of the limestone.
3. Oxidation of  $\text{FeO}$  in the slag by the furnace gases, the resulting oxide,  $\text{Fe}_2\text{O}_3$ , being reduced to  $\text{FeO}$  at the slag-metal surface.
4. Addition of iron ore to the charge, or to the bath during working.

### *Lime ( $\text{CaO}$ )*

Lime is the element which confers most of the basicity to the slag. The amount of lime in the slag is one of the determining factors in the extent of elimination of phosphorus and sulphur. The silica from the oxidation of silicon in the iron is neutralized by lime and excess lime is used in bringing about the elimination of phosphorus and sulphur.

### *Silica ( $\text{SiO}_2$ )*

For a given amount of lime charged to a furnace the amount of silicon in the iron governs the basicity of the slag, the erosion of the furnace lining, the condition of the slag after all the metal is melted and the extent of elimination of carbon by the time the heat is melted. Almost all of the silicon is eliminated in a short time after the metal is added, and in this reaction a large amount of the iron oxide in the slag is consumed. If the silicon in the iron is high, little iron oxide will be left for the removal of carbon and the heat will melt "high" in carbon.

### *Manganese Oxide ( $\text{MnO}$ )*

The chief function of manganese oxide is to increase the fluidity of the slag. If the concentration becomes too high the slag will be excessively "thin" and will cause trouble in working the heat. On the other hand, if the  $\text{MnO}$  is too low the slag will tend to become "heavy" and viscous and will require fluorspar, or some other thinning agent, to give it fluidity. The range of

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Si 1.17
Per Cent s



MnO in the slag for the best conditions of fluidity is from 7 to 9 per cent. It is not generally realized that the amount of dissolved MnO in the steel is proportional to the amount of MnO in the slag, and with certain classes of steel it may be desirable to keep the MnO in the slag low to prevent the nonmetallic inclusion MnO from being present in the steel in large enough quantities to injure the surface of the finished product.

### Magnesia ( $MgO$ )

Magnesia in the slag is derived chiefly from the erosion of the lining. Its action is probably to increase the viscosity of the slag, although no definite data is available on this point.

### COMPARISON OF HEATS

The three heats to be considered will be called A<sup>1</sup>, B, and C. A brief description of the essential factors will be given in the following paragraphs and in Table I.

*Furnace.* All these heats were made in standard 100-ton stationary furnaces of the Bethlehem Steel Co., Lackawanna, N. Y.

*Scrap.* The scrap in A consisted entirely of rail heat butts.

Table I

Heat	A			B			C		
CHARGE									
Per Cent Metal.....	31.6			36.0			64.2		
Per Cent Scrap .....	68.4			64.0			35.8		
TOTAL WEIGHT									
Pounds .....	255,540			242,000			240,000		
Ore .....	None			None			20,000		
LIMESTONE									
Pounds .....	26,000			28,000			28,000		
Per Cent of Tot. Wt. ....	10.2			11.6			11.7		
ORE USED IN WORKING									
Pounds .....	11,400			8,000			16,000		
BURNT LIME USED IN WORKING									
Pounds .....	None			None			3,000		
ANALYSIS OF CHARGE									
	Metal	Scrap	Ave.	Metal	Scrap	Ave.	Metal	Scrap	Ave.
C	4.14	0.67	1.77	4.29	0.66	1.96	4.04	0.20	2.66
Mn	1.37	0.84	1.01	1.28	0.72	0.92	1.29	0.30	0.94
P	0.36	0.025	0.132	0.21	0.028	0.093	0.27	0.020	0.189
S	0.059	0.033	0.041	0.039	0.037	0.038	0.052	0.045	0.049
Si	1.17	0.15	0.47	1.58	0.18	0.69	1.41	0.040	0.92
Per Cent Sulphur in Coal	1.18			.89			1.47		

<sup>1</sup>Keats, J. L., and Herty, C. H., Jr., *Transactions, A. I. M. M. E.*, February, 1926.

This gave a small ratio of surface area to volume, and would be classed as "large scrap." Heat B was charged as "medium scrap", with a fairly large ratio of surface area to volume. Heat C was charged with miscellaneous medium sized scrap and would fall in the same class as heat B.

*Fuel.* Each heat was made with producer gas. The sulphur content of the coal used is given in Table I.

These three heats will now be compared to bring out the salient features of open-hearth working. The points to be considered are:

1. Absorption of sulphur by the scrap during melting.
2. Oxidation of scrap during melting.
3. Slag formed after addition of the hot metal, and the elimination of impurities from the metal one hour after its addition.
4. Elimination during the lime boil and the composition of the slag at the end of the lime boil.
5. Action of fluorspar addition to heat B at end of lime boil.
6. The ore addition and its effect on slag and metal.
7. Final slag composition and its effect on the metal.
8. Final steel.

#### *Absorption of Sulphur by the Scrap During Melting*

Samples of the molten oxide which drips off the scrap during the melting period give the following average analyses:

Heat	Per Cent Sulphur in drippings	Per Cent Sulphur in coal
B .....	0.054	0.89
A .....	0.074	1.18
C .....	0.129	1.47

The total amount of sulphur absorbed during this period, for a given coal, is a function of the surface of scrap exposed and of the time required to melt. With small scrap and a high sulphur coal large amounts of sulphur will be absorbed. The faster the heat is melted the less will be the sulphur absorbed.

#### *Oxidation of the Scrap During Melting*

As has been shown previously the amount of iron oxide formed during melting strongly influences the amount of impurities which will be eliminated during the early stages of the heat.

Heat	Type of Scrap	Pounds of FeO formed per 100 pounds of Scrap	Time from charging scrap to hot metal addition
A .....	Large	1.76	2 hrs. 56 min.
B .....	Medium	8.35	2 hrs. 50 min.
C .....	Medium	8.33	3 hrs. 13 min.

The three heats are strictly comparable as to time of exposure for oxidation and the large difference in oxide formed from the two types of scrap is clearly shown. The effect of this will be seen in the comparison between heats A and B in the next section.

*Condition of the Bath 1 Hour after the Hot Metal Addition*

At this time the metal is boiling around the mounds of scrap and is also being agitated by the gases evolved from the decomposition of the limestone. The composition of slag and metal at this period on each heat is given below.

Table II

Heat	A	B	C
	Per Cent	Per Cent	Per Cent
Metal C .....	2.54	1.75	2.22
Mn .....	0.87	0.27	0.11
P .....	0.193	0.103	0.054
S .....	0.043	0.050	0.066
Si .....	0.394	0.073	.0065
Slag FeO .....	2.92	6.11	9.50
Fe <sub>2</sub> O <sub>3</sub> .....	0.39	0.13	1.14
SiO <sub>2</sub> .....	37.50	34.04	28.38
P <sub>2</sub> O <sub>5</sub> .....	0.18	0.50	3.64
MnO .....	15.18	12.38	11.20
S (Total) .....	0.255	0.160	0.157
CaO .....	33.62	34.25	33.70
MgO .....	4.85	9.50	5.88
Al <sub>2</sub> O <sub>3</sub> .....	4.17	....	....

The calculated slag weights are:

A	2,370 pounds
B	9,050 pounds
C	20,200 pounds

and this, combined with the iron oxide content of the slag, shows the effect of scrap oxidation on heats A and B and of the ore charged in heat C. As the oxidation of the slag increases, the amounts of carbon, manganese, phosphorus and silicon in the metal decrease, this being clearly shown in Table II. It is of particular interest to note that with a heavy oxidation of the slag the manganese will be eliminated to a low point, in heat C to a lower point than the tapping test.

The per cent carbon in heat C is higher than in A or B simply because there was a much larger amount of carbon in the charge on this heat. The amount of sulphur in the metal on heat A is low on account of the small absorption of sulphur from the gas, this being due to the small exposed area of the scrap, while in heat B the sulphur is higher, due in this case to the much larger area, the relative areas being shown roughly by the amount of iron oxidized. In heat C the sulphur in the metal is high due to the high sulphur coal used, to the size of scrap exposed to the furnace gases, and to the high sulphur in the charge.

#### ELIMINATION DURING THE LIME BOIL AND THE COMPOSITION OF THE SLAG AT THE END OF THE LIME BOIL

During the "lime boil" lime, resulting from the decomposition of the limestone charged, rises to the surface of the bath. The amount of lime dissolved by the slag during this period depends to a large extent on the composition and fluidity of the slag, and has a large effect on the elimination of phosphorus and sulphur. The silicon at this time is reduced to a small concentration. Furthermore, during this period the melted metal is being diluted by melting scrap which slowly changes the composition of the melted portion. The slag and metal composition at the end of the lime boil is shown in Table III.

Table III

		A Per Cent	B Per Cent	C Per Cent
Metal	C .....	1.16	1.04	1.60
	Mn .....	0.56	0.38	0.22
	P .....	0.105	0.070	0.040
	S .....	0.029	0.043	0.054
Slag	FeO .....	3.89	3.45	6.79
	Fe <sub>2</sub> O <sub>3</sub> .....	1.72	0.94	1.80
	SiO <sub>2</sub> .....	28.48	31.78	28.18
	P <sub>2</sub> O <sub>5</sub> .....	1.45	0.55	3.02
	MnO .....	15.44	10.66	9.64
	S (Total) .....	0.047	0.143	0.164
	CaO .....	37.90	38.95	39.55
	MgO .....	7.28	9.74	6.18
	Al <sub>2</sub> O <sub>3</sub> .....	2.99	....	....

The first observation here is that the iron oxides in heats A and B are about the same whereas heat C is still higher in

these oxides. The carbon in all three heats has dropped considerably due to dilution and elimination. Manganese has returned to the metal on heats B and C where the concentrations of iron oxides in the slag have decreased and has been eliminated slightly in heat A. In all three heats both phosphorus and sulphur concentrations have decreased, the phosphorus content of the metal again being the reverse of the oxidation of the slag, heat C versus heats A and B, and of the total amount of phosphorus in the charge, heats A and B. Sulphur in all three heats has been eliminated.

#### ACTION OF FLUORSPAR ADDITION TO HEAT B AT END OF LIME BOIL

At the end of the lime boil in heat B 700 pounds of fluorspar were added to "shape up" the slag. This thinned out the slag and caused a large amount of lime to be dissolved, with a consequent drop in phosphorus and sulphur content. The essential analyses just before and  $\frac{1}{2}$  hour after the fluorspar addition are given in Table IV.

Table IV

		Before Fluorspar Addition	$\frac{1}{2}$ hour after Fluorspar Addition
		Per Cent	Per Cent
Metal	C .....	1.04	0.73
	Mn .....	0.38	0.46
	P .....	0.070	0.021
	S .....	0.043	0.035
Slag	FeO .....	3.45	4.92
	Fe <sub>2</sub> O <sub>3</sub> .....	0.94	1.17
	SiO <sub>2</sub> .....	31.78	20.41
	CaO .....	38.95	51.83

The solution of lime with a correspondingly large rise in the per cent CaO in the slag served to remove phosphorus and sulphur rapidly. The increased fluidity of the slag speeded up the oxidation of the slag and carbon was eliminated rapidly. Manganese is still returning to the metal as was the case in Table III.

#### ACTION OF ORE

When ore is added all the oxidizable elements are diminished. Sulphur, on the other hand, is usually returned to the metal from the slag, due to the increased oxidation of the slag. The action of ore is shown in Table V where analyses just before and  $\frac{1}{2}$  hour after the addition are given.



Table V

Heat	A		B		C	
Pounds of Ore added .....	7,000		8,000		4,000	
	A		B		C	
Metal	Before Per Cent	After Per Cent	Before Per Cent	After Per Cent	Before Per Cent	After Per Cent
C .....	1.16	0.92	0.73	0.32	1.60	1.30
Mn .....	0.56	0.22	0.46	0.16	0.22	0.16
P .....	0.105	0.045	0.021	0.007	0.040	0.015
S .....	0.029	0.034	0.035	0.036	0.054	0.057
Slag						
FeO ....	3.89	14.48	4.92	11.69	6.79	9.61
Fe <sub>2</sub> O <sub>3</sub> ...	1.72	2.68	1.17	5.81	1.80	3.62

## FINAL SLAG COMPOSITION AND ITS EFFECT ON THE METAL

During the working period the slag is kept oxidized by the furnace gases and, where necessary, by suitable additions of ore. Thus in heats A and C ore was added during this period, whereas sufficient ore had been added to heat B at the first addition and no more was used. Lime floating in the slag is dissolved and the banks of the furnace considerably eroded during this period, the basicity of the slag being controlled for a given silicon in the iron by these factors.

The elimination of all the metalloids is brought to the desired point and the heat is then ready to be tapped. It should be mentioned that it is usually true that in working the carbon down to the desired point, phosphorus is eliminated much further than called for by the specifications on the steel. This is not so often the case with sulphur since the iron oxide which eliminates carbon tends to retard the elimination of sulphur. Thus with fairly low phosphorus materials charged this element may be disregarded as far as its elimination is concerned.

The important points to be kept in mind regarding the working period are:

1. That the slag must be sufficiently oxidized to eliminate carbon at a fairly rapid rate, but not so highly oxidized that the elimination of sulphur is slowed up.
2. That the slag must be basic enough to eliminate sulphur to the desired point.
3. That the slag be sufficiently fluid to allow elimination to proceed rapidly.

Very often the slag may be of the correct composition to bring

the sulphur as low as is desired, but the fluidity is not high enough to allow the elimination to take place without keeping the heat in the furnace unduly long. This point is beautifully illustrated by the removal of sulphur in heat C during the working period as shown in Table VI. The slag composition during the period shown in the table changed but little.

Table VI

Time	Per Cent Sulphur in Steel	Condition of Slag
6:30 .....	0.057	Very viscous
7:10 .....	0.057	Very viscous
7:10 .....	.....	Began adding fluorspar
7:30 .....	0.057	Thinning
7:45 .....	0.051	Fluid
8:00 .....	0.047	Fluid
8:32 .....	0.044	Fluid
9:08 .....	0.043	Fluid

It is in this question of slag fluidity that the high manganese iron is so important, although it must be remembered that all slags high in MnO are not necessarily fluid. For the average tapping or working slags, however, increasing manganese increases the fluidity greatly. Fluorspar is widely used as a thinning agent, and unless used in excessive amounts is satisfactory. The relative action of two slags, thinned respectively by MnO and by fluorspar, on the cleanliness of the steel has never been studied and at present the use of either is primarily a question of cost and of the inclination of the person using these materials.

The final metal and slag analyses, before the final additions were made, are given in Table VII.

Table VII

Heat No.		A	B	C
		Per Cent	Per Cent	Per Cent
Metal	C .....	0.22	0.09	0.07
	Mn .....	0.24	0.17	0.11
	P .....	0.010	0.009	0.004
	S .....	0.023	0.025	0.043
	FeO .....	11.37	11.85	15.54
Slag	Fe <sub>2</sub> O <sub>3</sub> .....	3.12	7.10	2.49
	SiO <sub>2</sub> .....	15.86	14.48	19.14
	P <sub>2</sub> O <sub>5</sub> .....	3.67	2.32	2.56
	MnO .....	9.11	6.63	6.58
	S (Total) .....	0.157	0.278	0.186
	CaO .....	47.65	47.45	42.90
	MgO .....	6.00	6.20	6.10
	Al <sub>2</sub> O <sub>3</sub> .....	2.26	.....	.....
	Final Temp. Degree Fahr.	2895	2980	2840

In Table VII the action of the oxidation and the basicity

of the slag on the elimination of phosphorus and sulphur is clearly shown. It is a well known fact that phosphorus may be more completely removed, for a given slag composition, at a low temperature than at a high one. Of these three heats, heat C has the lowest percentage phosphorus on account of the low temperature and high oxidation of the slag, and in spite of the lowest basicity. Heats A and B are about equal basicity (heat B being slightly higher) and the higher oxidation of B is counter-balanced by its higher temperature. Sulphur, on the other hand, is little affected by temperature alone, and the sulphur content of A and B are practically the same, the slightly higher basicity of B being counter-balanced by its higher oxidation. Heat C has the highest sulphur content due to both high oxidation and low basicity.

The manganese content of the metal follows the manganese content of the slag, the temperature and the oxidation of the slag, the factors giving low residual manganese being:

- 1 Low MnO in the slag
- 2 High oxidation of the slag
- 3 Low temperature.

Heat A was tapped while the carbon was dropping fairly rapidly. The carbon content on both heats B and C was decreasing very slowly at the end of the heat and they have both reached approximately the same content with about equal oxidation of the slag.

#### FINAL STEEL

The additions made to the furnace or ladle in the three heats, and the composition of the final steel is given in Table VIII.

Table VIII				
Additions	Heat	A <sup>1</sup>	B <sup>2</sup>	C <sup>3</sup>
	Ferromanganese .....	1500	1500	1100
	Ferrosilicon .....	700	....	....
	Coal .....	420	....	....
Analysis of Steel		Per Cent	Per Cent	Per Cent
	C .....	0.372	0.12	0.09
	Mn .....	0.608	0.45	0.35
	P .....	0.014	0.013	0.006
	S .....	0.021	0.039	0.045
	Si .....	0.081	....	....

<sup>1</sup>Average of analysis from ingots Nos. 2, 5, 9, 13, 17 and 21.

<sup>2</sup>Average of analysis from ingots Nos. 2, 12, 27.

<sup>3</sup>Analysis of ingots No. 2 only.

## CARBON

The increase in carbon over the tapping test, Table VII, is due in heats B and C to the carbon contained in the ferromanganese.

## MANGANESE

The efficiencies of the manganese additions were—

A—79 per cent,      B—56 per cent,      C—66 per cent

In the addition of manganese four factors affect the efficiency of the addition. First, the oxidation of the metal, which for a given temperature, is proportional to the oxidation of the slag, and which increases with rising temperature. Second, temperature, which affects the oxidation of the metal as shown above, and which, as it increases, causes an increase in reaction velocity. Third, time allowed for the deoxidizing reaction to take place. Fourth, addition of other deoxidizers.

In the three heats the time between addition of manganese and end of teeming was equal enough to be neglected in comparing the efficiencies of the additions. The effect of the other three factors is shown below.

Table IX

Efficiency of Manganese Addition

Heat	A	B	C
Efficiency .....	79	56	66
Oxidation of Slag ( $\text{FeO} + \text{Fe}_2\text{O}_3$ )...	14.49	18.95	18.03
Temp. Degrees Fahr. ....	2895	2980	2840
Other deoxidizers .....	Yes	No	No

The effect of temperature is shown in the comparison of heats B and C, while heat A which had a temperature between those of B and C gave the highest efficiency due to the low oxidation of the slag and to the presence of other deoxidizers.

## Phosphorus

Phosphorus returns to the metal due to the deoxidation of the slag by the deoxidizers used and to the decrease in basicity of the slag by the erosion of the ladle lining. A small "kick-back" of phosphorus is shown in all three heats.

### *Sulphur*

The action of sulphur in the ladle is not well understood. It can be stated, however, that the hotter the heat and the lower the basicity of the slag the greater the sulphur reversion, the higher the manganese and the greater the deoxidation of the slag the smaller the sulphur reversion.

### CONCLUSION

The author prefers to limit the paper to the open-hearth furnace itself rather than discuss the effect of the process on the finished steel. Before this effect can be intelligently studied it is necessary to thoroughly understand the process itself and it is the aim of this paper to point out the essential features of the whole, leaving for other more detailed papers, the exposition of the separate details.

### DISCUSSION

DR. HAAKON STYRI: Just a few words regarding the statement on page 573 where Dr. Herty says: "The amount of dissolved MnO in the steel is proportional to the amount of MnO in the slag." I think that should be elaborated upon a little, because it must depend on the other constituents of the slag. If you have comparatively much silica present with a certain percentage of manganese oxide in the slag you will have less manganese oxide in the steel than if you have high lime content in the slag and low silica, so the chemical composition of the slag is certainly of importance regarding the amount of manganese oxide dissolved.

DR. C. H. HERTY: I can very easily change that to suit Dr. Styri and say for a given slag composition the MnO in the steel is proportional to the MnO in the slag.



## THE DECOMPOSITION OF THE AUSTENITIC STRUCTURE IN STEELS—Part III.\*

The Effect of Drawing or Tempering on the Decomposition of Austenite—Normal Stresses

BY RALPH L. DOWDELL AND OSCAR E. HARDER

### *Abstract*

*In this series of experiments the effect of tempering quenched steels for different periods of time at different temperatures has been studied by means of the changes in hardness and the changes in microstructures. The steels included in this investigation are the same as those reported in Part I, The Decomposition of Austenite During Quenching.*

*In this series of tests some attention has also been given to the effect of severe mechanical stress, such as up-setting an austenitic steel, on its decomposition on tempering. It has been found that austenite more frequently tempers to troostite, but in certain cases typical martensitic needles have been produced on tempering. The temperature at which the martensitic structure breaks down in any given steel has been found to be lower than the temperature at which the austenitic structure in the same steel breaks down. There is found a general agreement in the changes in the microstructure, the hardness, and specific gravity, and explanations are offered for the mechanism of these changes.*

THE tempering (drawing) tests to be discussed in the following pages have been made in order to show the breakdown of the austenitic structure of quenched steels on reheating at various temperatures. The changes in hardness and microstructure have been used as the principal means in following this decomposition. In addition to the tempering tests certain specimens have been tempered in the stressed condition in order to determine the influence of a combination of stress and reheating on the changes in quenched steels.

\*This and five other articles on The Decomposition of the Austenite Structure in Steel are based on a manuscript submitted by Ralph Lewis Dowdell in partial fulfillment of the requirements of the Graduate School of the University of Minnesota for the degree of Doctor of Philosophy. O. E. Harder, in charge of research. These articles have appeared serially in TRANSACTIONS since January, 1927. Seven installments constitute this series.

A paper presented before the eighth annual convention of the Society, Chicago, September 20 to 24, 1926. Of the authors, Dr. R. L. Dowdell is assistant professor of metallography and Dr. O. E. Harder is professor of metallography, University of Minnesota, Minneapolis.

*General Description*—The tempering experiments were made on the several steels by placing them in an ordinary automatically controlled oven for long periods of time. This oven was found to be accurate to plus or minus about 3 degrees Cent. up to its maximum temperature. These long periods of time at the tempering temperatures were chosen in order to study the physical changes more closely and to more nearly obtain equilibrium conditions.

When it was found that the tempering reaction became too slow at 392 degrees Fahr. (200 degrees Cent.) the steels were tempered in an electric tube furnace provided with variable resistances which enabled quite an accurate method of temperature control. The temperatures up to 392 degrees Fahr. (200 degrees Cent.) were measured with a mercury thermometer while the higher ones were measured with a chromel-alumel couple.

*Tests on Bain's High Carbon-Chromium Steel*—Because of the interest in E. C. Bain's<sup>1</sup> researches on chromium steels, a special study was made of one of his steels. Table VI shows the effect of cumulative low tempering temperatures for various periods of time on the scleroscope hardness of specimens of steel No. 3 quenched from temperatures from 1500 to 2100 degrees Fahr. (815 to 1149 degrees Cent.).

**Table VI**  
**Effect of Time of Tempering at 100° C. and at 200° C. on the Scleroscope Hardness of Steel No. 3**

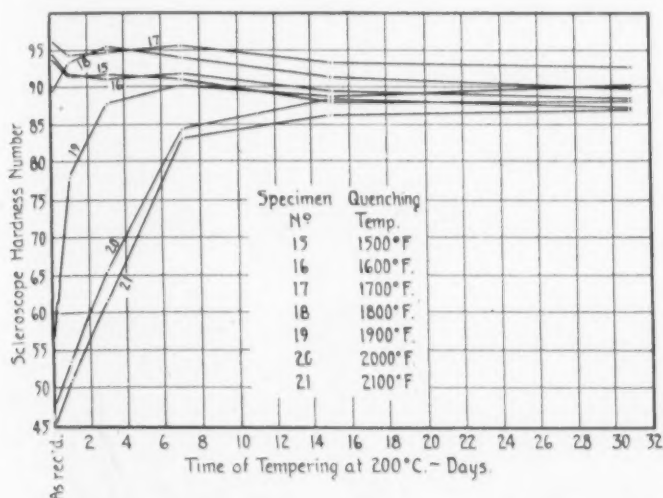
Specimen Numbers	15	16	17	18	19	20	21
Oil Quenching Temp. °F. 1500	1600	1700	1800	1900	2000	2100	
Oil Quenching Temp. °C. 815	871	927	982	1038	1093	1149	
No. Heat Treatment	Scleroscope Hardness Numbers						
Original	94.3	94.7	96.5	89.4	55.0	47.4	44.9
1 24 hrs. at 100° C.	94.4	94.7	99.7	92.7	58.0	46.8	46.7
2 72 hrs. at 100° C.	97.7	95.8	100.8	94.8	59.9	50.0	48.4
3 163 hrs. at 100° C.	94.3	93.5	99.7	92.2	60.5	48.5	48.5
4 168 hrs. at 100° + 24 hrs. at 200° C.	91.7	91.9	94.5	93.6	78.1	53.6	51.3
5 168 hrs. at 100° + 72 hrs. at 200° C.	91.8	91.2	94.9	95.5	88.0	65.8	61.3
6 168 hrs. at 100° + 168 hrs. at 200° C.	91.2	92.1	95.7	94.1	90.4	84.6	83.3
7 168 hrs. at 100° + 360 hrs. at 200° C.	88.1	89.7	93.5	91.6	88.9	88.5	86.4
8 168 hrs. at 100° + 31 days at 200° C.	88.5	88.8	93.1	90.1	90.5	87.6	86.9

Note: Scleroscope values are the averages of ten readings (5 on each side of the  $\frac{3}{8}$ " square specimen).

<sup>1</sup>TRANSACTIONS, American Society for Steel Treating, Vol. 5, 1924, pp. 89-101.

These results are shown on Graph No. 1. On examination of this graph it will be noticed that the austenitic specimens, those quenched from 1900 degrees Fahr. (1038 degrees Cent.) and above, show a rapid increase in hardness and finally come to have essentially the same hardness as the other specimens quenched from a lower temperature.

Figs. 44 to 62 inclusive show the austenitic decomposition on tempering the quenched specimens shown in the previous figures Nos. 8 to 14 inclusive. Only the austenitic specimens have

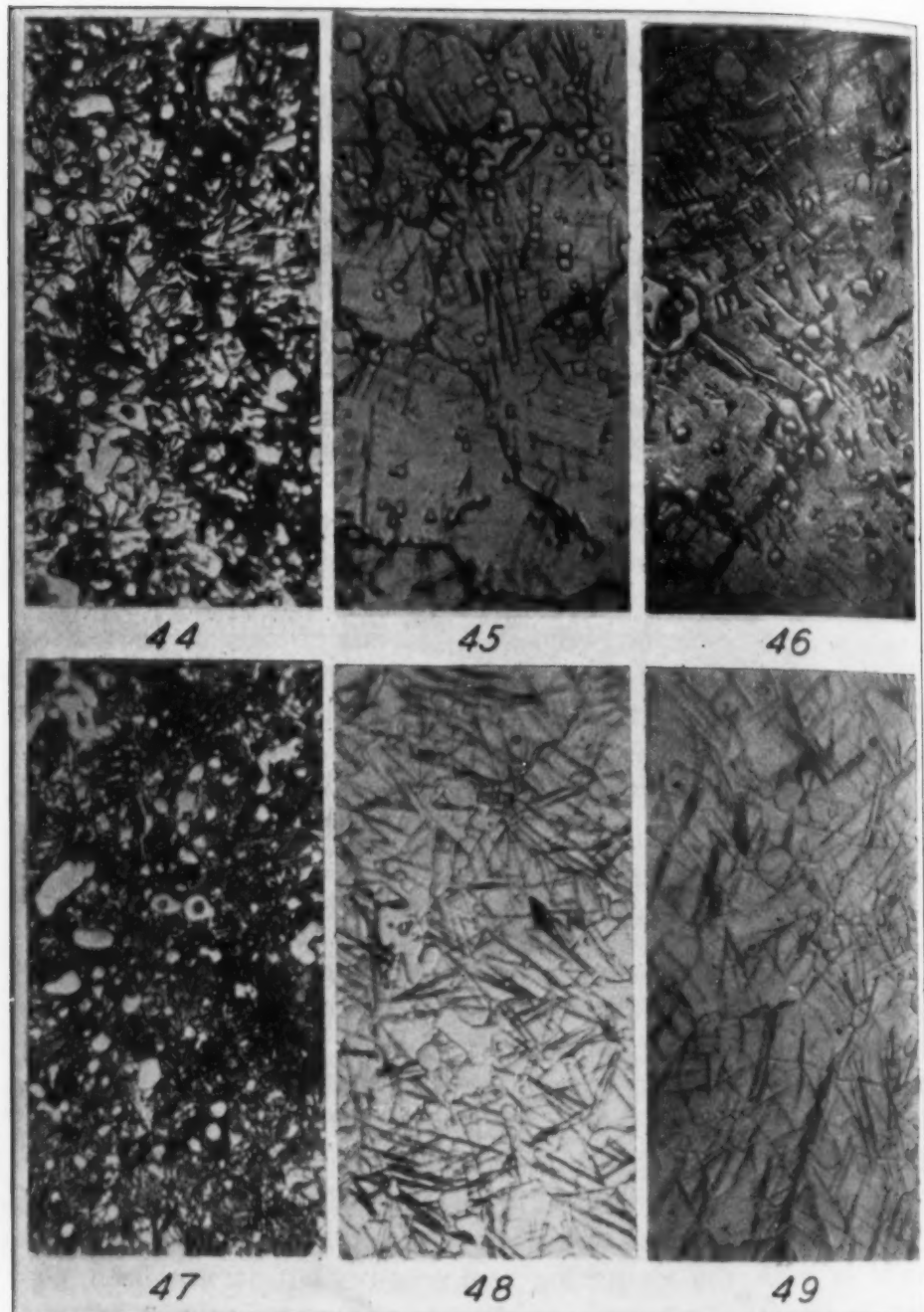


Graph 1—Effect of Time of Tempering at 392 Degrees Fahr. on the Hardness of No. 3 High Chromium Steel.

been examined throughout the tests. These photomicrographs show that the recrystallization started at the slip planes in the austenite. Instead of forming martensite on this recrystallization it is quite probable that the recrystallization starts and is followed almost immediately by precipitation of carbide to form troostite. However, it was found quite impossible to form white needles on tempering this steel by the method used.

The time required to break down the austenitic structure increases with the quenching temperature but it has been possible to completely decompose the austenitic structure at 392 degrees Fahr. (200 degrees Cent.) and after the 744-hour tempering there is indication of the coalescence of the carbide particles.

*Effect of Tempering on Microstructure and Hardness*—Figs. 63 to 109 inclusive show the results of tempering on the microstructures of steels Nos. 1 to 6 inclusive. Table VII shows the effect



Effect of Tempering Temperatures on the Decomposition of the Austenitic Structure Produced by Quenching from 1500, 1600, 1700, 1800, 1900, 2000 and 2100 Degrees Fahr. Steel No. 3. Figs. 44, 45 and 46. Tempering Treatment No. 4, 168 Hours at 212 Degrees Fahr. + 24 Hours at 392 Degrees Fahr. Fig. 44—Specimen No. 19. Mag. 1000x. Scleroscope Reading after = 78.1; before = 55.0. Etched with Conc. HCl. Austenite + Troostite + Carbide. Fig. 45—Specimen No. 20. Mag. 1000x. Scleroscope Reading after = 53.6; before = 47.4. Etched with Conc. HCl. Austenite + Troostite + Carbide. Fig. 46—Specimen No. 21. Mag. 1000x. Scleroscope Reading after = 51.3; before = 44.9. Etched with Conc. HCl. Austenite + Troostite + Carbide. Figs. 47, 48 and 49. Tempering Treatment No. 5, 168 Hours at 212 Degrees Fahr. + 72 Hours at 392 Degrees Fahr. Fig. 47—Specimen No. 19. Mag. 1000x. Scleroscope Reading after = 88.0; before = 55.0. Etched with Conc. HCl. Troostite + Carbide + Austenite. Fig. 48—Specimen No. 20. Mag. 1000x. Scleroscope Reading after = 65.8; before = 47.4. Etched with Conc. HCl. Austenite + Troostite + Carbide. Fig. 49—Specimen No. 21. Mag. 1000x. Scleroscope Reading after = 61.8; before = 44.9. Etched with Conc. HCl. Austenite + Troostite + Carbide.

Fig.  
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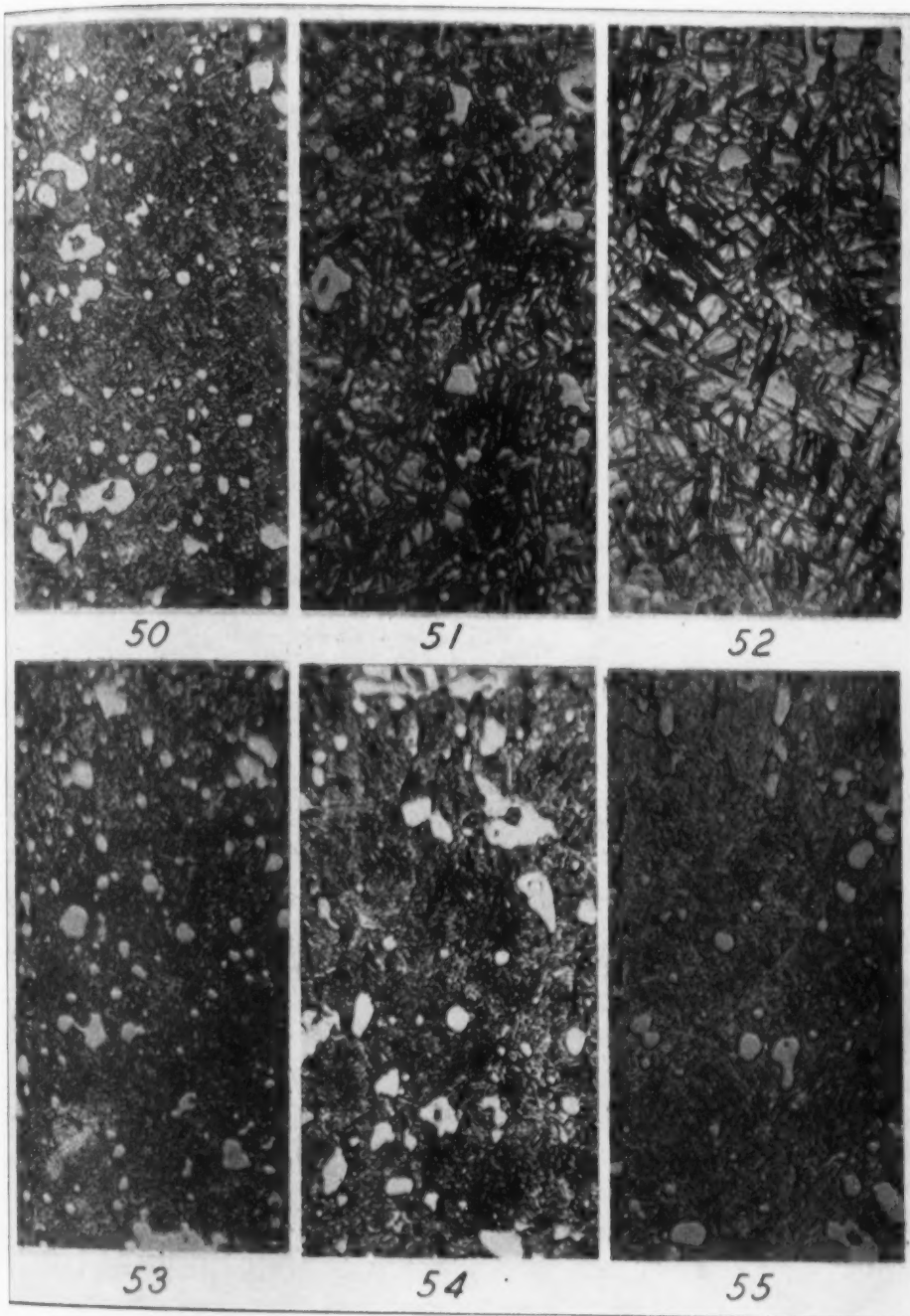
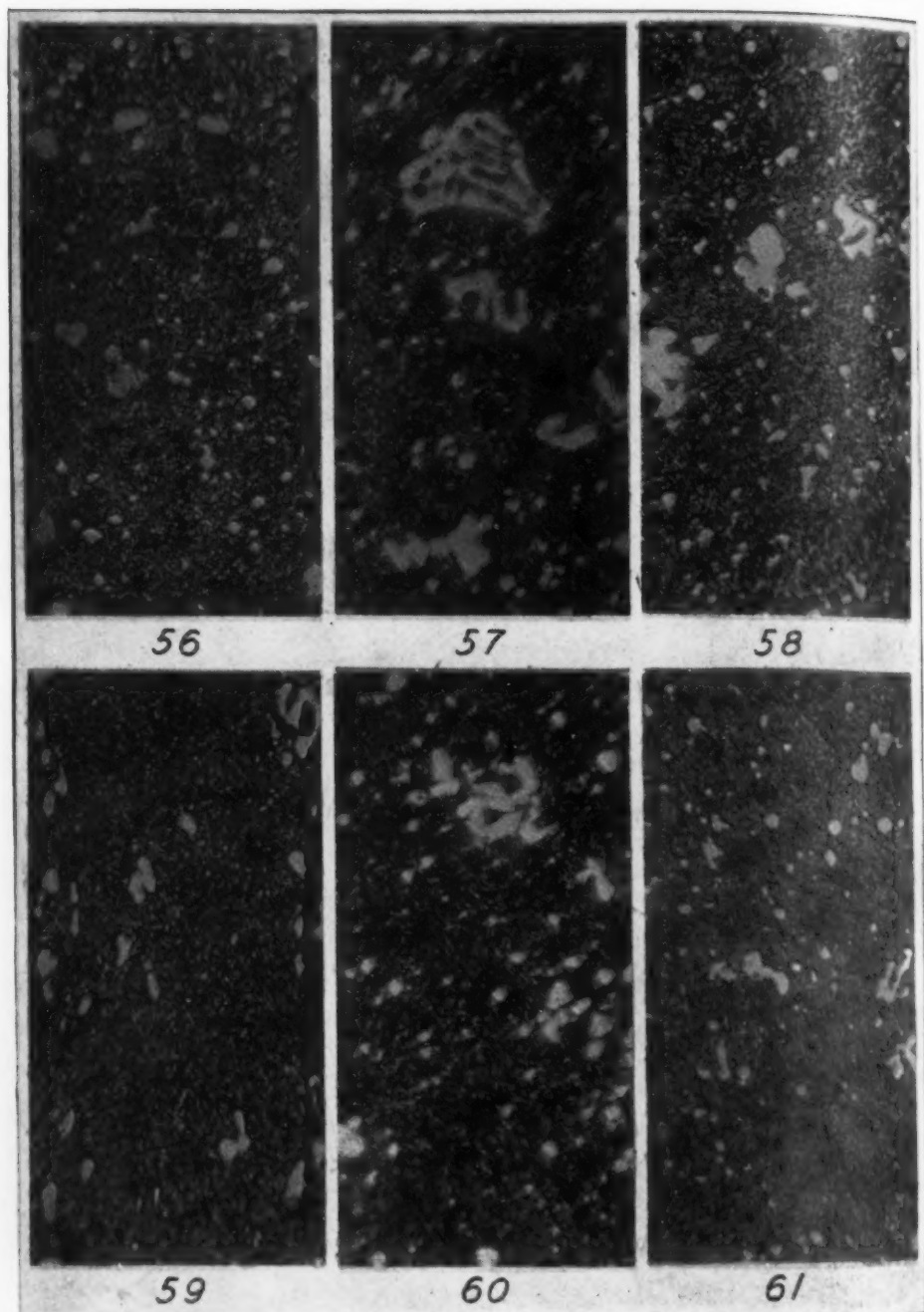


Fig. 50 to 52. Tempering Treatment No. 6, 168 Hours at 212 Degrees Fahr. + 168 Hours at 392 Degrees Fahr. Fig. 50—Specimen No. 19, Mag. 1000x. Scleroscope Reading after = 90.4; before = 55.9. Etched with Conc. HCl. Troostite + Carbide. Fig. 51—Specimen No. 20, Mag. 1000x. Scleroscope Reading after = 84.6; before = 47.4. Etched with Conc. HCl. Austenite + Troostite + Carbide. Fig. 52—Specimen No. 21, Mag. 1000x. Scleroscope Reading after = 83.3; before = 44.9. Etched with Conc. HCl. Austenite + Troostite + Carbide. Figs. 53 to 55. Tempering Treatment No. 7, 168 Hours at 212 Degrees Fahr. + 360 Hours at 392 Degrees Fahr. Fig. 53—Specimen No. 19, Mag. 1000x. Scleroscope Reading after = 88.9; before = 55.0. Etched with Conc. HCl. Troostite + Carbide. Fig. 54—Specimen No. 20, Mag. 1000x. Scleroscope Reading after = 88.5; before = 47.4. Etched with Conc. HCl. Troostite + Carbide. Fig. 55—Specimen No. 21, Mag. 1000x. Scleroscope Reading after = 86.4; before = 44.9. Etched with Conc. HCl. Troostite + Carbide.





Figs. 56 to 62. Steel No. 3. Tempering Treatment No. 8, 168 Hours at 212 Degrees Fahr. + 744 Hours at 392 Degrees Fahr. Fig. 56—Specimen No. 15. Mag. 1000x. Scleroscope Reading after = 88.5; before = 94.3. Etched with Conc. HCl. Troostite + Carbide. Fig. 57—Specimen No. 16. Mag. 1000x. Scleroscope Reading after = 88.8; before = 94.7. Etched with Conc. HCl. Troostite + Carbide. Fig. 58—Specimen No. 17—Mag. 1000x. Scleroscope Reading after = 93.1; before = 89.3. Etched with Conc. HCl. Troostite + Carbide. Fig. 60—Specimen No. 19. Mag. 1000x. Scleroscope Reading after = 90.5; before = 55.0. Etched with Conc. HCl. Troostite + Carbide. Fig. 61—Specimen No. 20. Mag. 1000x. Scleroscope Reading after = 87.6; before = 47.4. Etched with Conc. HCl. Troostite + Carbide.

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of tempering for various periods of time and at various temperatures on the austenitic decomposition of the steels mentioned above.

Table VII

Effect of Tempering Temperatures on Scleroscope Hardness Values During the Austenite Breakdown of Steels Nos. 1 to 6

Number of Heat Treatment	1	2	3	4	5	6
Size of Specimen Treated	.26" d	.44" x .69" sq.	.385" sq.	.49" sq.	.40" d	.45" d
Original As Hardened	45.98	27.7	46.4	53.6	46.4	79.7
A 1 day at 100° C. (212° F.)	....	30.4	....	....	....	85.2
B 2 days at 100° C. (212° F.)	....	31.1	....	....	....	82.3
C 4 days at 100° C. (212° F.)	46.0	28.7	45.9	56.0	46.7	89.5
D 8 days at 100° C. (212° F.)	47.3	29.2	46.0	54.5	46.3	86.0
E 2 days at 200° C. (392° F.)	49.8	30.2	52.7	62.5	48.1	90.4
F 4 days at 200° C. (392° F.)	46.5	32.5	84.6	84.3	47.2	91.1
G 8 days at 200° C. (392° F.)	46.6	33.7	86.9	90.4	48.0	....
H 16 days at 200° C. (392° F.)	49.8	30.2	92.4	93.2	48.5	90.5
I 1 day at 260° C. (500° F.)	49.5	31.3	90.3	95.5	50.3	91.5
J 3 days at 270° C. (518° F.)	52.1	33.3	92.2	94.5	50.5	89.1
K 3 days at 318° C. (607° F.)	51.3	33.2	92.0	91.2	50.6	83.3
L 1 day at 388° C. (730° F.)	49.9	33.3	87.5	93.2	55.6	73.2
M 1 day at 410° C. (770° F.)	51.4	41.0	86.6	91.2	50.5	68.4
N 3 days at 400-410° C.	51.6	47.4	87.0	91.0	50.2	68.6
O 7 days at 410° C.	52.4	52.9	87.0	91.9	49.1	66.8
P 7 days at 450° C. (842° F.)	54.7	66.6	81.0	92.3	46.1	67.0
Q 7 days at 480° C. (896° F.)	53.8	61.4	77.1	90.0	50.1	51.7
R 7 days at 525° C. (979° F.)	82.8	59.0	71.9	86.6	46.3	46.7
S 7 days at 562° C. (1044° F.)	78.6	55.6	64.9	85.8	42.5	39.9
T 7 days at 625° C. (1158° F.)	66.4	52.0	53.9	69.5	38.3	25.8
U 7 days at 643° C. (1192° F.)	64.8	51.2	51.9	63.8	36.1	24.5
V 6 days at 670° C. (1240° F.)	62.0	49.8	54.3	65.9	36.6	27.5

Notes: Table III shows the hardening treatment.

S. h. n. values are averages of 5 readings.

E, No. 6 was only 1 day at 200°.

F, No. 6 was only 2 days at 200°.

O, No. 2 is the only one with any noticeable change in microstructure.

Q, No. 5 is a new specimen.

U, No. 6, new specimen hardened = 25.5.

V, No. 6, new hardened specimen.

*Steel No. 1*—The cobalt-chromium steel No. 1 used in this series of tests was essentially austenitic as quenched. Careful tests were made by different methods of etching to see if it was possible to bring out martensite needles. The hardness as quenched (46) also indicates quite a thoroughly austenitic structure. After it was tempered 3 days at 518 degrees Fahr. (270 degrees Cent.) the hardness increased to 52 and the first indication of martensite needles was found as shown in Fig. 64. These needles were not general but occurred only in certain areas. After the specimen was tempered for one week at 979 degrees Fahr. (525 degrees Cent.) the needles became much more numerous and were general all over the specimen. The structure as developed by

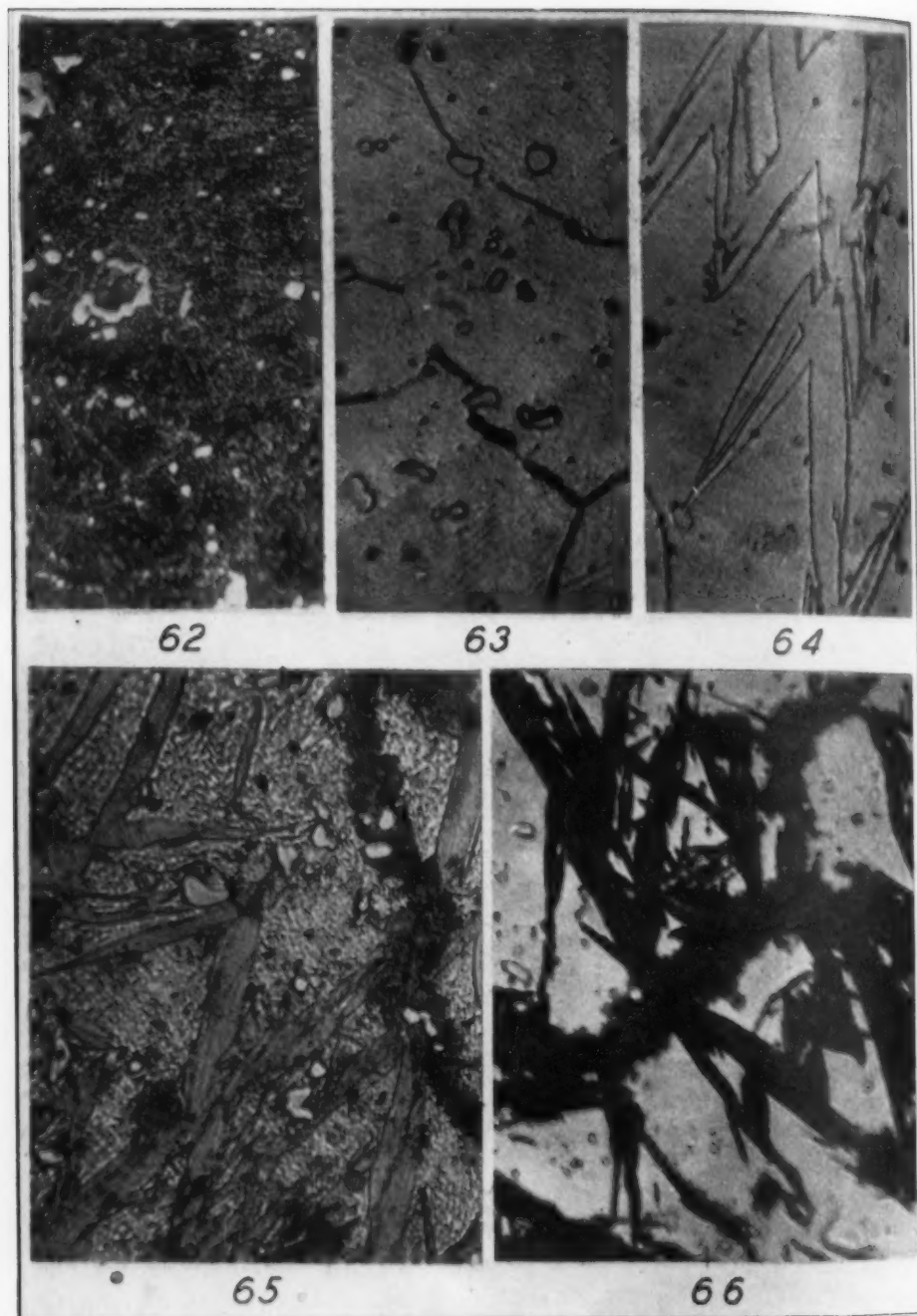


Fig. 62—Steel No. 3. Tempering Treatment No. 8, 168 Hours at 212 Degrees Fahr. + 744 Hours at 392 Degrees Fahr. Fig. 62—Specimen No. 21. Mag. 1000x. Scleroscope Reading after = 86.9; before = 44.9. Etched with Conc. HCl. Troostite + Carbide. Decomposition of Austenite on Tempering. Steel No. 1. Cobalt + Chromium. Quenching Treatment 2282 Degrees Fahr. 15 minutes 5 per cent NaOH in Water at 25 Degrees Fahr. Fig. 63—Mag. 1000x. After Quench. Scleroscope Reading = 46.0. Etched with HCl. Austenite + Carbide. Fig. 64—Mag. 1000x. Tempered 3 days at 518 Degrees Fahr. Scleroscope Reading = 52.1. Etched with Conc. HCl. Only Needles on Specimen. Austenite + Martensite + Carbide. Fig. 65—Mag. 1000x. Tempered 7 days at 977 Degrees Fahr. Scleroscope Reading = 82.8. Etched with Conc. HCl. Austenite + Troostite + Carbide. Fig. 66—Mag. 1000x. Tempered 7 days at 977 Degrees Fahr. Scleroscope Reading = 82.8. Etched with Picric Acid. Austenite + Troostite + Carbide.

Fig.  
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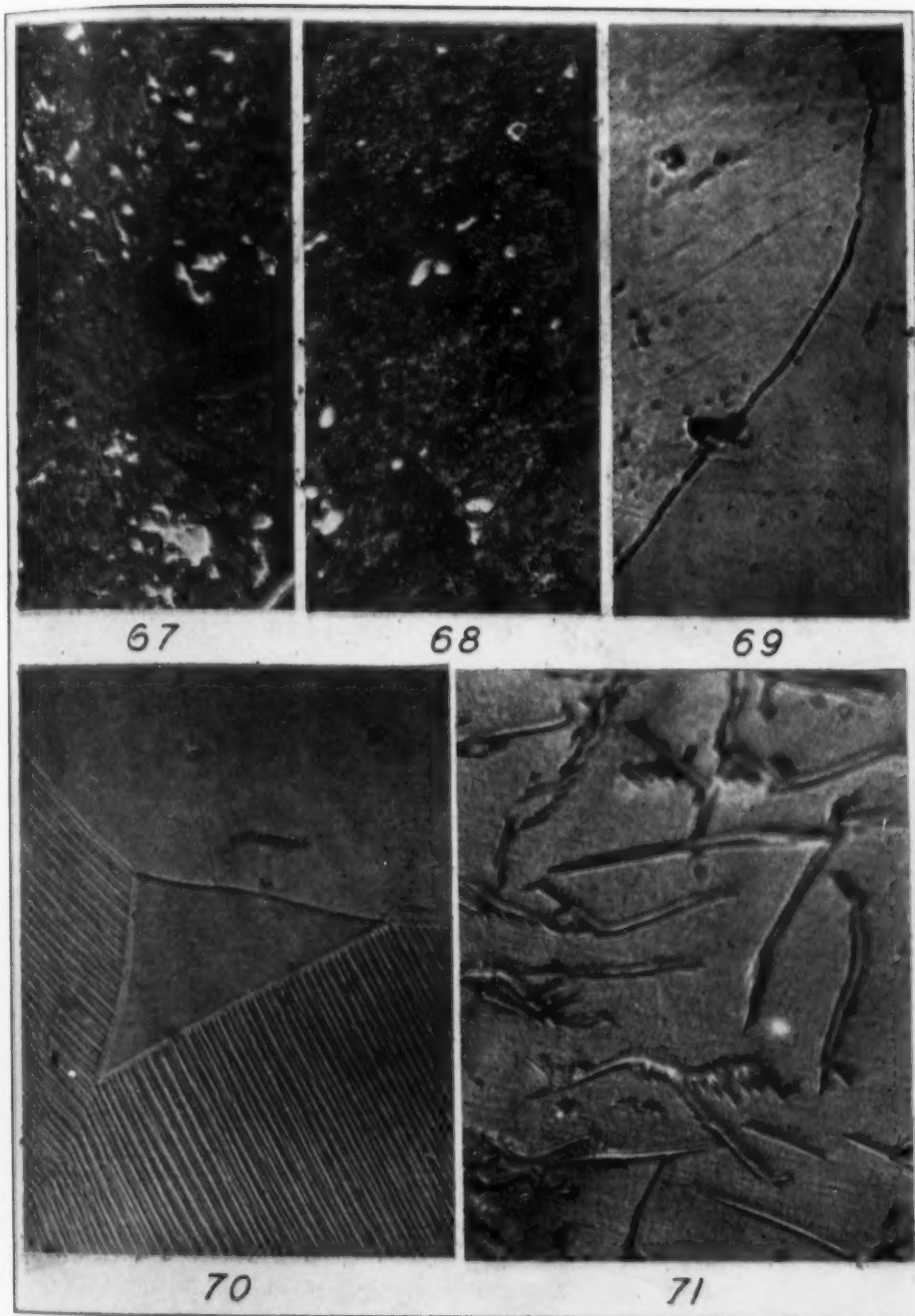


Fig. 67—Mag. 1000x. Tempered 7 days at 1045 Degrees Fahr. Scleroscope Reading = 78.6. Etched with Picric Acid. Troostite + Carbide. Fig. 68—Mag. 1000x. Tempered 7 days at 1157 Degrees Fahr. Scleroscope Reading = 66.4. Etched with Picric Acid. Troostite + Carbide.

Figs. 69 to 78. Decomposition of Austenite on Tempering. Steel No. 2, Hadfield's Manganese Quenching Treatment 2102 Degrees Fahr. 15 minutes, 5 per cent NaOH in Water at 25 Degrees Fahr. Fig. 69—Mag. 1000x. After Quench. Scleroscope Reading = 27.7. Etched with Picric Acid. Austenite. Fig. 70—Mag. 1000x. After Quench, more Deeply Etched. Scleroscope Reading = 27.7. Etched with Picric Acid. Austenite with Slip lines. Fig. 71—Mag. 1000x. Tempered 1 day at 770 Degrees Fahr. Scleroscope = 41.0. Etched with Picric Acid. Austenite + Martensite.



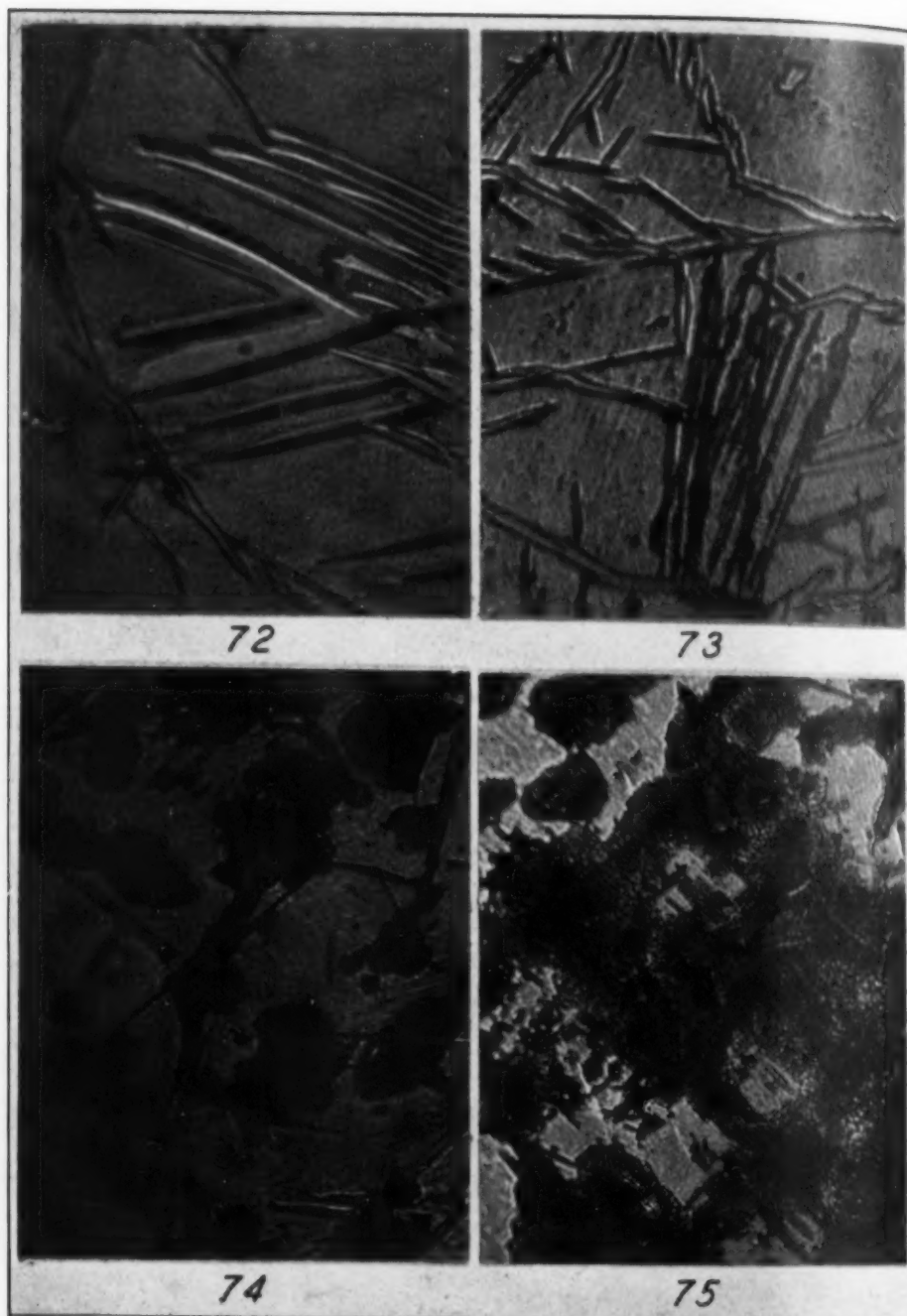


Fig. 72—Mag. 1000x. Tempered 1 Day at 770 Degrees Fahr. Another Field to Show Needles. Etched with Picric Acid. Austenite + Martensite. Fig. 73—Mag. 1000x. Tempered 7 days at 770 Degrees Fahr. Scleroscope Reading = 52.9. Etched with Picric Acid. Austenite + Martensite. Fig. 74—Mag. 1000x. Tempered 7 days at 842 Degrees Fahr. Scleroscope Reading = 66.6. Etched with Picric Acid. Austenite + decomposing Martensite + Troostite. Fig. 75—Mag. 1000x. Tempered 7 days at 896 Degrees Fahr. Scleroscope = 61.4. Etched with Picric Acid. Austenite + Troostite-Sorbite.



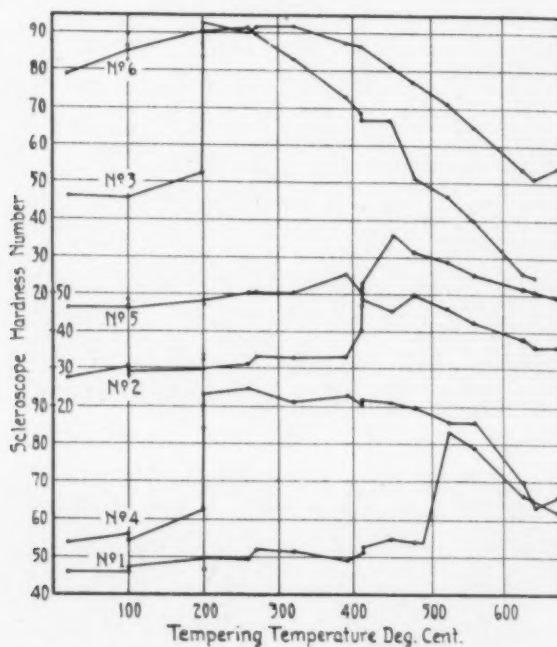
hydrochloric acid etching is shown in Fig. 65 where the needles are lighter than when etched in picric acid (Fig. 66). In both etches the presence of the austenitic groundmass is clearly shown and the needles are darker in color than the groundmass. The hardness after this tempering treatment had increased to 83. On tempering one week at 1158 degrees Fahr. (625 degrees Cent.) the needles completely disappeared and the troostite was much coarser (Fig. 68). This resulted in a hardness to only 66.

However, there seems to be an overlapping of the tempering temperature which produces the maximum amount of these needles and the temperature at which they begin to temper. Thus after 3 days at 518 degrees Fahr. (270 degrees Cent.) relatively few needles were produced and they showed little or no tempering. There was a large increase in the number of needles when the tempering treatment was carried to 979 degrees Fahr. (525 degrees Cent.) but they begin to show tempering. The changes in the hardness of this steel can best be followed by reference to Graph No. 2.

*Steel No. 2*—High manganese steel No. 2 as quenched showed a typical austenitic structure (Fig. 69); on deep etching many slip lines were observed. On tempering steel No. 2 it was noticed that after one day's tempering at about 770 degrees Fahr. (410 degrees Cent.), there was a marked increase in hardness due to the formation of white needles which were in relief (Figs. 71 and 72). These needles are probably martensitic needles although they are slightly unusual in appearance. On tempering for one week at 770 degrees Fahr. (410 degrees Cent.) more needles were formed which etched light and resulted in an increase of hardness over the original hardness by about 76 per cent. After tempering for one week at 842 degrees Fahr. (450 degrees Cent.) it will be noticed (Fig. 74) that the needles had darkened and also that the remaining austenitic groundmass had begun to break down to form a nodular troostite. The maximum hardness was produced with this tempering treatment and resulted in an increase of about 112 per cent. Continued tempering at a higher temperature resulted in a decrease of hardness. The microstructure produced (Fig. 75) indicates the coalescence of carbide particles. There are, however, still areas of undecomposed austenite and slight remnants of these are found even after tempering

for one week at 1040 degrees Fahr. (560 degrees Cent.). The changes in hardness of this steel for the various tempering temperatures are best shown in Graph No. 2.

The light needles produced in this steel were produced on tempering and remained light in color to quite an elevated tem-



Graph 2—Effect of Tempering Temperature on the Scleroscope Hardness of Steels Nos. 1 to 6 During Their Austenitic Breakdown.

perature. This may mean this austenite is under considerable compression or that its recrystallizing temperature on tempering is high, due to its greater stability caused by the high manganese content.

*Steel No. 3*—Further tempering experiments on steel No. 3 on a sample which was requenched by heating to 2192 degrees Fahr. (1200 degrees Cent.), held for 15 minutes and quenched in a 5 per cent solution of NaOH in water at 24 degrees Fahr. (−4 degrees Cent.), are in general agreement with those previously discussed. In no case were typical martensite needles produced on tempering.

A small amount of martensite was found in the extreme edge of the quenched specimen (Fig. 80). This may indicate that the extreme edge was in the proper amount of tension during the quenching and so produced martensite. Complex stresses such

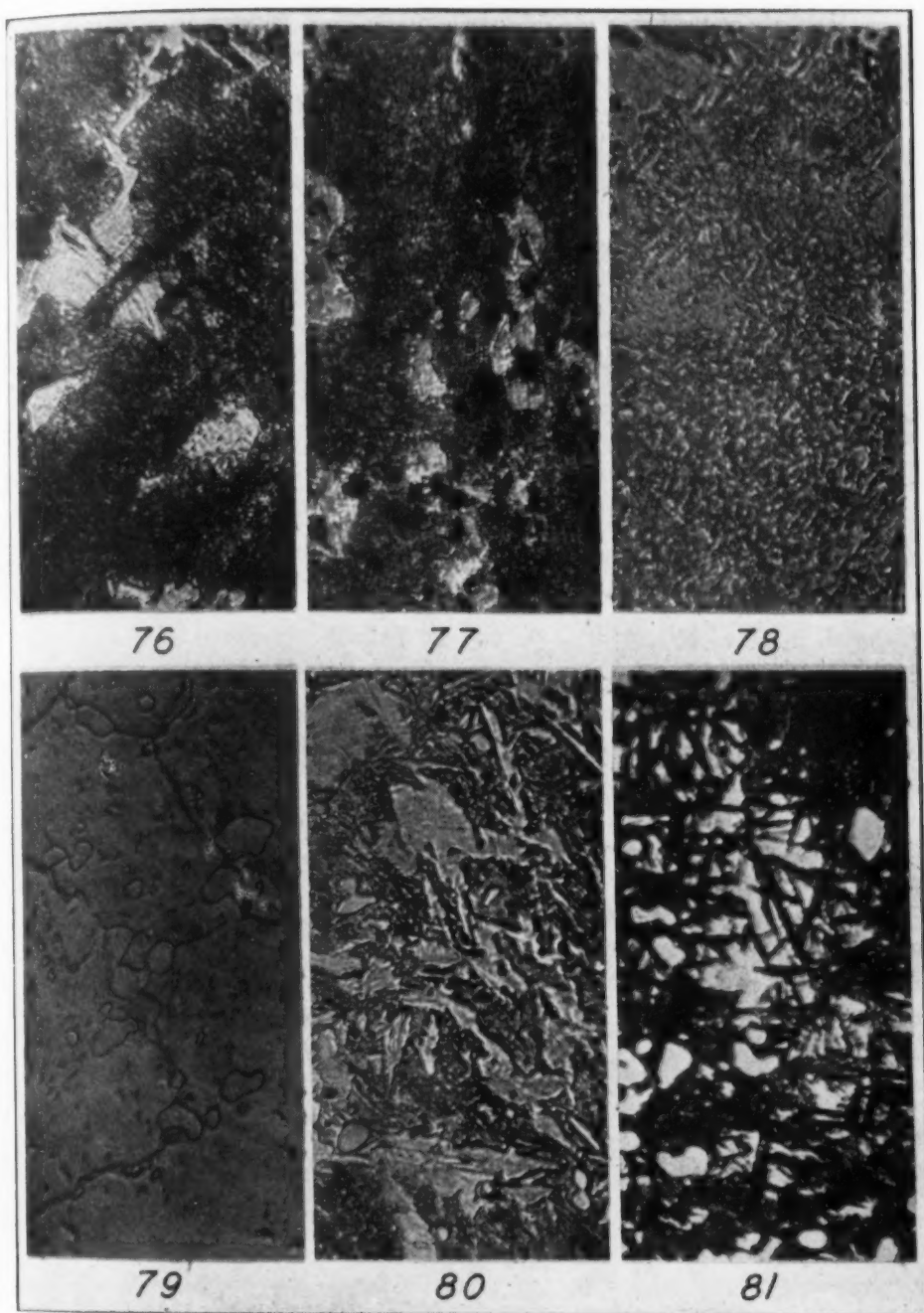


Fig. 76—Mag. 1000x. Tempered 7 days at 979 Degrees Fahr. Scleroscope = 59.0. Etched with Picric Acid. Austenite + Troostite-Sorbite. Fig. 77—Mag. 1000x. Tempered 7 days at 1044 Degrees Fahr. Scleroscope = 55.6. Etched with Picric Acid. Austenite + Troostite-Sorbite. Fig. 78—Mag. 1000x. Tempered 7 days at 1158 Degrees Fahr. Scleroscope = 52.0. Etched with Picric Acid. Sorbite + Ferrite. Figs. 79, 80, 81, 82, 83, 84, 85. Decomposition of Austenite on Tempering. Steel No. 3 High Carbon-Chromium. Quenching Treatment 2192 Degrees Fahr. 15 minutes, 5 per cent NaOH in Water at 25 Degrees Fahr. Fig. 79—Mag. 1000x. After Quench. Scleroscope Reading = 46.4. Etched with Conc. HCl. Austenite + Carbide. Fig. 80—Mag. 1000x. After Quench. About .5 millimeter from outer edge. Etched with HCl. Austenite + Martensite + Carbide. Fig. 81—Mag. 1000x. Tempered at 392 Degrees Fahr. Scleroscope Reading = 84.6. Etched with Conc. HCl. Austenite + Troostite + Carbide.

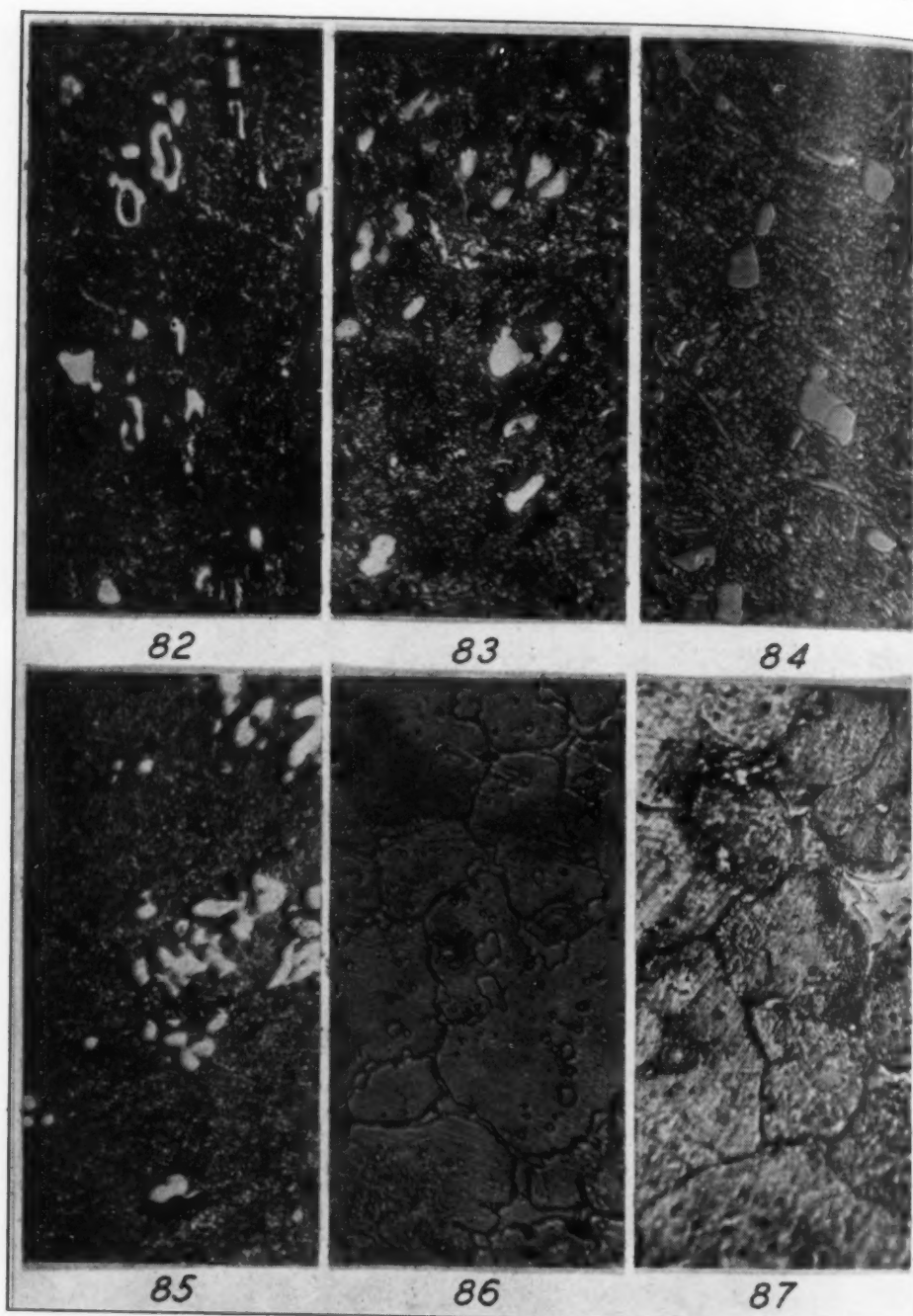


Fig. 82—Mag. 1000x. Tempered 3 days at 518 Degrees Fahr. Scleroscope Reading = 92.2. Etched with Conc. HCl. Troostite + Carbide. Fig. 83—Mag. 1000x. Tempered 7 days at 979 Degrees Fahr. Scleroscope Reading = 71.9. Etched with Conc. HCl. Troostite + Carbide. Fig. 84—Mag. 1000x. Tempered 7 days at 1158 Degrees Fahr. Scleroscope Reading = 53.9. Etched with Conc. HCl. Sorbite + Carbide. Fig. 85—Mag. 1000x. Tempered 4 days at 374 Degrees Fahr. Etched with Conc. HCl. Troostite + Carbide. Fig. 86—Steel No. 4 Mag. 1000x. After Quench. Scleroscope Reading = 53.6. Etched with Picric Acid. Austenite + Carbide. Fig. 87—Mag. 1000x. Tempered 2 days at 392 Degrees Fahr. Scleroscope Reading = 62.5. Etched with Picric Acid. Austenite + Troostite + Carbide.



as are produced by hammering as in upsetting a steel of this type materially lower the temperature at which this steel breaks down to a troostite on tempering. A specimen which was upset so as to increase the scleroscope hardness from 47 to 85 was changed entirely to a troostitic structure by a 4-day tempering treatment at 374 degrees Fahr. (190 degrees Cent.).

*Steel No. 4*—The specimen of high speed steel used in the tempering test was heated to 2282 degrees Fahr. (1250 degrees Cent.), held at temperature 15 minutes and quenched until cold in the caustic solution. This steel decomposed on tempering similarly to steel No. 3 and white needles could not be found at any stage in this tempering treatment. The breakdown of the austenite to troostite took place at first principally at the grain boundaries (Fig. 87). On tempering two days at 392 degrees Fahr. (200 degrees Cent.) the hardness increases only slightly but with additional time at 200 degrees Cent. there was a considerable increase in hardness accompanied by a corresponding change in microstructure (Figs. 88 and 89 and Graph No. 2). While the decomposition of the austenite after eight days' tempering at 392 degrees Fahr. (200 degrees Cent.) was found to be practically complete, the hardness was found to increase slightly at 518 degrees Fahr. (270 degrees Cent.) and the groundmass appeared to be somewhat coarser (Fig. 90). At about 1158 degrees Fahr. (625 degrees Cent.) coalescence of the carbide particles took place rapidly (Fig. 91) which resulted in a marked decrease in hardness.

*Steel No. 5*—The progress of tempering steel No. 5 is shown in Figs. 92 to 96. This austenite remained stable throughout the cumulative tempering tests to a temperature as high as 1240 degrees Fahr. (670 degrees Cent.) without appreciable decomposition. At about 1158 degrees Fahr. (625 degrees Cent.) it was noticed that some of the carbide patches began to decompose into ferrite and graphite. The photomicrographs also indicate that graphite forms directly from the austenite on tempering at 1158 degrees Fahr. (625 degrees Cent.) and above. However, when graphitization takes place there is a marked decrease in hardness which is best shown in Graph No. 2.

*Steel No. 6*—The progress of the decomposition of steel No. 6 is shown in Figs. 97 to 109 inclusive. On examination of Fig.

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- Carbide.  
Fig. 93.9.  
4 days at  
No. 4 Mag.  
Austenite  
Scleroscope



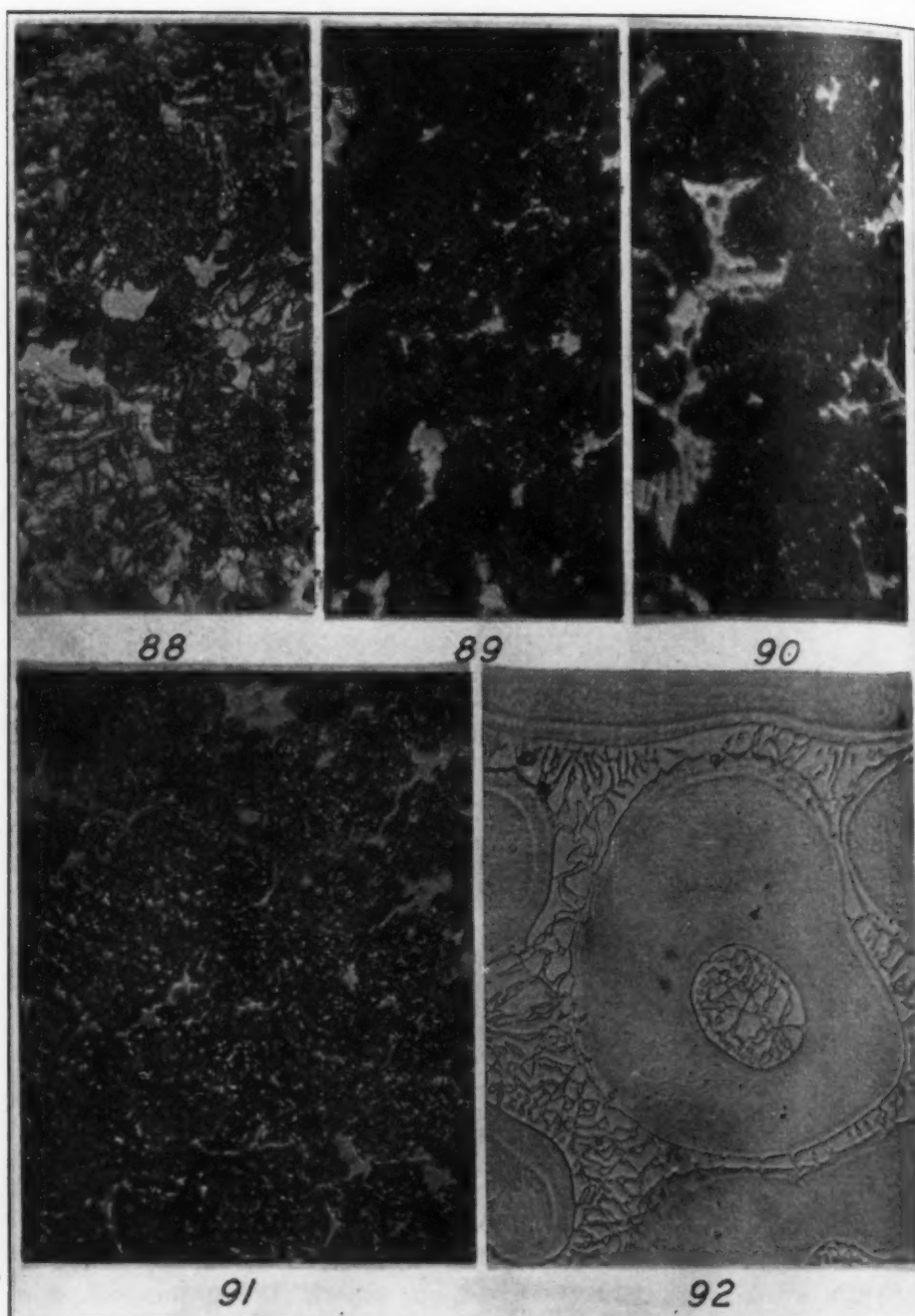
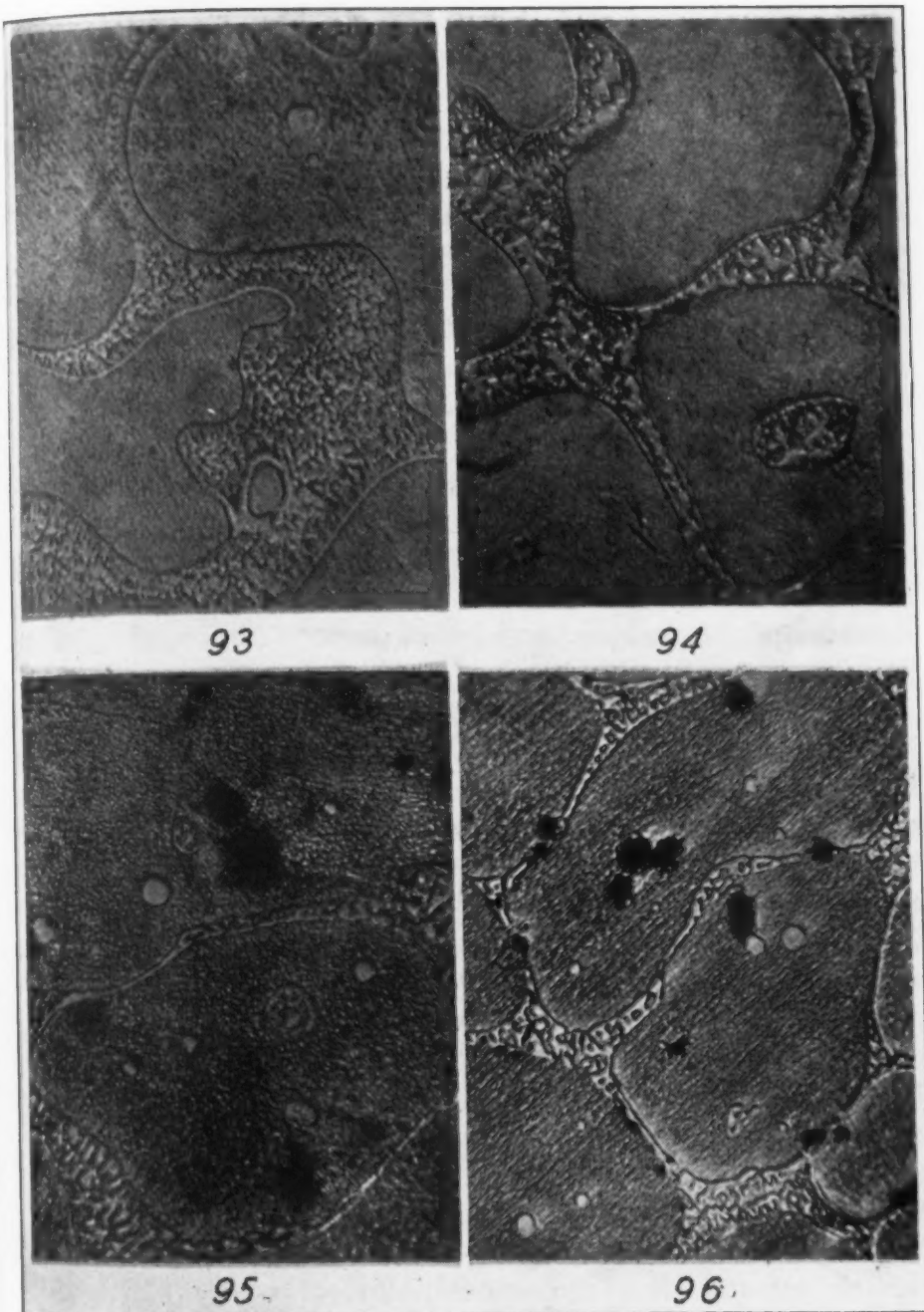


Fig. 88—Mag. 1000x. Tempered 4 days at 392 Degrees Fahr. Scleroscope Reading = 84.3. Etched with Picric Acid. Troostite + Austenite + Carbide. Fig. 89—Mag. 1000x. Tempered 8 days at 392 Degrees Fahr. Scleroscope = 90.4. Etched with Picric Acid. Troostite + Carbide. Fig. 90—Mag. 1000x. Tempered 3 days at 518 Degrees Fahr. Scleroscope Reading = 94.5. Etched with Picric Acid. Troostite + Carbide. Fig. 91—Mag. 1000x. Tempered 7 days at 1158 Degrees Fahr. Scleroscope Reading = 69.5. Etched with Picric Acid. Sorbite + Carbide. Figs. 92, 93, 94, 95, 96. Decomposition of Austenite on Tempering. Steel No. 5, 25 per cent Nickel. Quenching Treatment 2280 Degrees Fahr. 15 minutes. 5 per cent NaOH in Water at 25 Degrees Fahr. Fig. 92—Mag. 1000x. After Quench. Scleroscope Reading = 46.4. Etched with Picric Acid. Austenite + Carbide.

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Figs. 93 to 96. Decomposition of Austenite on Tempering Steel No. 5, 25 per cent Nickel. Quenching Treatment 2280 Degrees Fahr. 15 minutes, 5 per cent NaOH in Water at 25 Degrees Fahr. Fig. 93—Mag. 1000x. Tempered 3 days at 518 Degrees Fahr. Scleroscope Reading = 50.5. Etched with Picric Acid. Austenite + Carbide. Fig. 94—Mag. 1000x. After Quench. New Specimen. Scleroscope Reading = 50.1. Etched with Picric Acid. Austenite + Carbide. Fig. 95—Mag. 1000x. Tempered 7 days at 1158 Degrees Fahr. Scleroscope Reading = 38.3. Etched with Picric Acid. Graphitization has taken place on edges of small Carbides. Austenite + Carbide + Graphite. Fig. 96—Mag. 1000x. Tempered 6 days at 1240 Degrees Fahr. Scleroscope Reading = 36.6. Etched with Picric Acid. Austenite + Carbide + Graphite.

97 it will be noted that the martensite needles after the quench are slightly darker than the austenite. Figs. 98 and 99 show the microstructures of this steel after tempering at 212 degrees Fahr. (100 degrees Cent.) for 4 and 8 days respectively. The martensite needles first darken and then further break down as the tempering progresses. The hardness shows a slight increase after a tempering treatment of 4 days but falls off slightly on further tempering. The amount of austenite is practically unchanged during the tempering treatment at 100 degrees Cent. The effects of the tempering treatment at 392 degrees Fahr. (200 degrees Cent.) are shown in Figs. 100 and 101. The breaking down of the austenite after tempering for one day is evident and after two days the breakdown is practically complete resulting in a marked increase in hardness. At the higher tempering treatments at 518 degrees Fahr. (270 degrees Cent.) (Fig. 102) and 730 degrees Fahr. (388 degrees Cent.) (Fig. 103), there is a coalescence of the carbide particles which is accompanied by a progressive decrease in hardness. At about 1040 degrees Fahr. (560 degrees Cent.) these small carbide particles began to decompose into ferrite and graphite which resulted in a further decrease in hardness. This graphitization is quite complete after annealing at about 1192 degrees Fahr. (643 degrees Cent.) for one week. See Figs. 104, 105 and 106. To check this phenomenon two new quenched pieces were tempered for a week at 1192 degrees Fahr. (643 degrees Cent.) and at about 1238 degrees Fahr. (670 degrees Cent.) respectively. The one at 1192 degrees Fahr. (643 degrees Cent.) showed a more complete graphitization than the one at 1238 degrees Fahr. (670 degrees Cent.) (Figs. 107 and 108). This is likely due to the fact that the carbide particles are larger and more stable when tempered at 1238 degrees Fahr. (670 degrees Cent.) and these larger particles are not as susceptible to graphitization as the smaller ones. Tempering tests on an annealed sample of this steel (Fig. 5; Fig. 109) bear out this theory and show that on still larger carbide particles the graphitization is still more incomplete.

*Steel No. 7*—Two differential heating curves were taken on quenched pieces of steel No. 7 (Fig. 110) to determine if it were possible to record a temperature effect attributable to the precipitation of carbide or recrystallization from martensite. No

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evolutions of heat were encountered on these curves. On microscopic examination it was found that the martensite had formed troostite but the austenite still remained undecomposed even after heating slowly to 852 degrees Fahr. (455 degrees Cent.).

*Steels Nos. 8 and 10*—On tempering steels Nos. 8 and 10 at 212 degrees Fahr. (100 degrees Cent.) it was found that the white martensitic needles formed during the quench were darkened considerably. This probably indicates that carbide precipitation takes place at this temperature. Microscopic evidence bears this out and shows that as the time at 212 degrees Fahr. (100 degrees Cent.) increases, the white needles (martensitic) assume a gradual increase in darkness.

The summary of results and conclusions of these tests are given later with the results from the studies of gravity changes during tempering and the effects of stress.

*Change in Specific Gravity*—Specific gravity determinations were made on oil and water quenched specimens of steel No. 8 and a water quenched specimen of steel No. 10, both quenched from 2012 degrees Fahr. (1100 degrees Cent.). The tests were made on the specimens in the quenched condition and after tempering for various periods of time at 212 degrees Fahr. (100 degrees Cent.) and at 374 degrees Fahr. (190 degrees Cent.). Table VIII shows the values obtained.

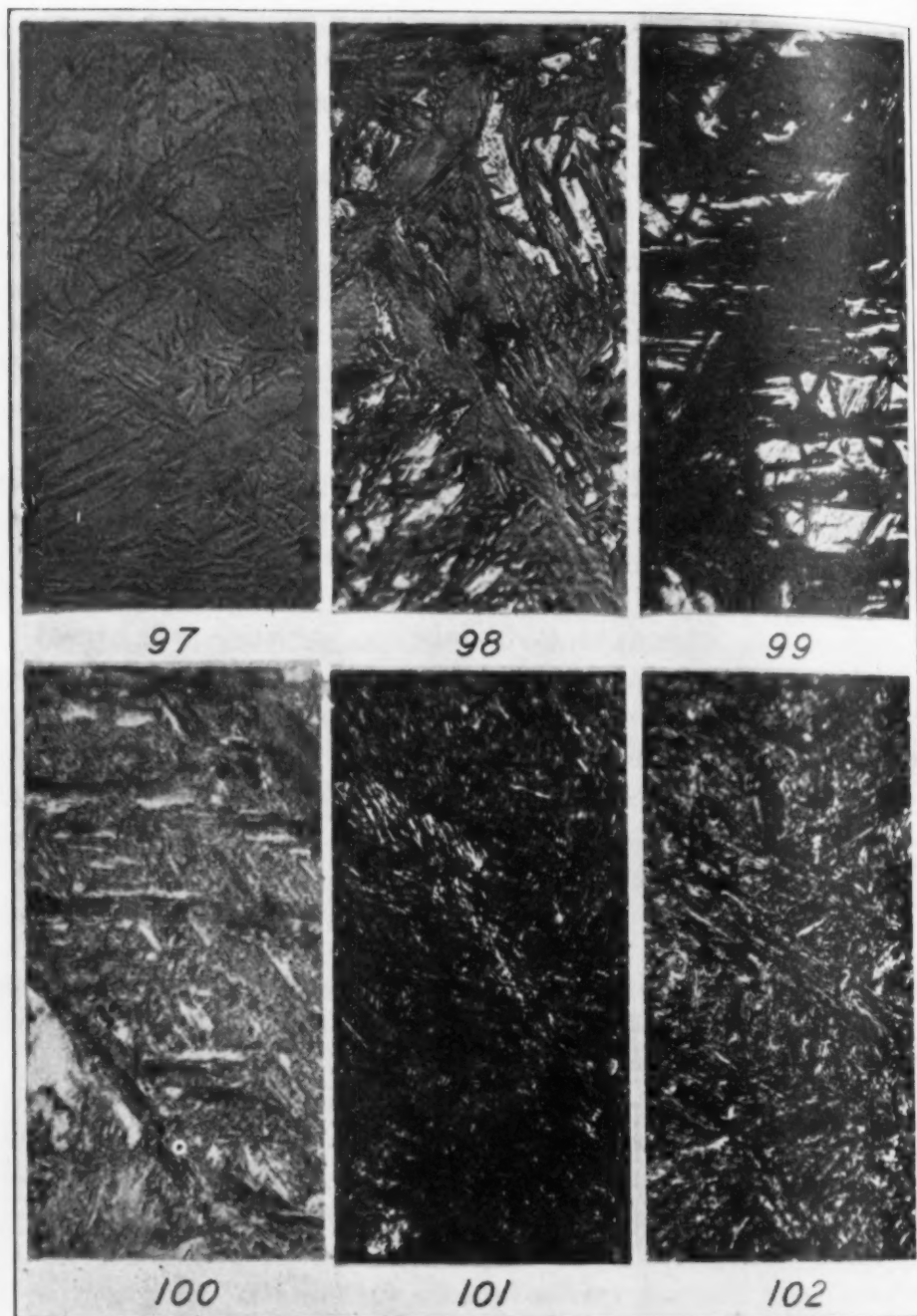
Table VIII

Effect of Tempering on the Specific Gravity and Scleroscope Hardness of Steels Nos. 8 and 10. Quenched from 2012° F. (1100° C.)

	Steel No. 8. 3/8"		Steel No. 8. 3/8"		Steel No. 10. 1"	
	Oil quenched.		Water quenched.		Water quenched.	
	S. h. n.	Sp. gr.	S. h. n.	Sp. gr.	S. h. n.	Sp. gr.
After hardening	82.2	8.0889	92.3	7.9648	73.3	7.6311
24 hrs. at 100° C.	84.7	8.0987	95.4	7.9971	78.4	7.6406
72 hrs. at 100° C.	84.5	8.1186	93.7	8.0263	71.6	7.6459
216 hrs. at 100° C.	86.6	8.1193	90.0	8.0089	74.2	7.6386
504 hrs. at 100° C.	84.8	8.1144	91.6	8.0386	76.0	7.6434
1 day at 190° C.	92.8	8.0816	96.0	7.9861	93.7	7.6021
4 days at 190° C.	99.1	8.0399	96.2	8.0135	97.7	7.5942
18 days at 190° C.	95.8	8.0542	93.8	8.0124	93.1	7.6066
48 days at 190° C.	95.2	8.0577	95.0	8.0789	91.9	7.6107

It was found that the density of steel No. 8 was greater in the oil-quenched condition than in the water-quenched which indicated more martensite on water quenching and more austenite after oil quenching.

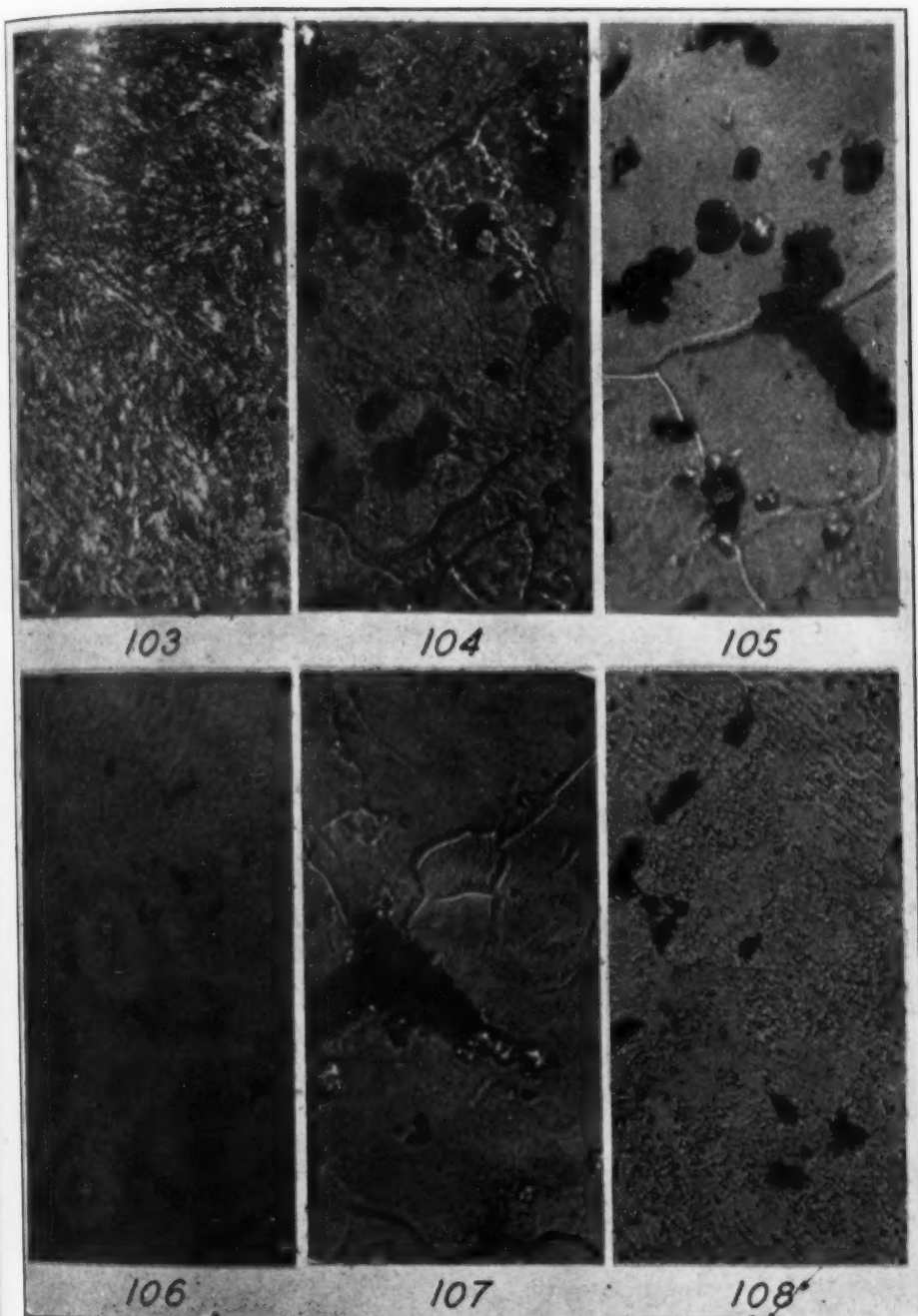




Figs. 97 to 109. Decomposition of Austenite on Tempering. Steel No. 6. High Carbon. 1.35 per cent Carbon. Quenching Treatment 2102 Degrees Fahr. 15 minutes, 5 per cent NaOH in Water at 25 Degrees Fahr. Fig. 97—Mag. 1000x. After Quench. Scleroscope Reading = 79.7. Etched with Picric Acid. Martensite + Austenite. Fig. 98—Mag. 1000x. Tempered 4 days at 212 Degrees Fahr. Scleroscope Reading = 89.5. Etched with Picric Acid. Troostite + Austenite. Fig. 99—Mag. 1000x. Tempered 8 days at 212 Degrees Fahr. Etched with Picric Acid. Troostite + Austenite. Fig. 100—Mag. 1000x. Tempered 1 day at 392 Degrees Fahr. Etched with Picric Acid. Troostite + Austenite. Fig. 101—Mag. 1000x. Tempered 2 days at 392 Degrees Fahr. Scleroscope Reading = 90.4. Etched with Picric Acid. Troostite. Fig. 102—Mag. 1000x. Tempered 3 days at 518 Degrees Fahr. Scleroscope Reading = 89.1. Etched with Picric Acid. Troostite.

Figs.  
Carbon, 1  
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730 Degree  
Fig. 104—  
25.8. Etc  
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Figs. 103 to 109. Decomposition of Austenite on Tempering. Steel No. 6. High Carbon, 1.35 per cent Carbon. Quenching Treatment 2102 Degrees Fahr. 15 minutes, 5 per cent NaOH in Water at 25 Degrees Fahr. Fig. 103—Mag. 1000x. Tempered 1 day at 730 Degrees Fahr. Scleroscope Reading = 73.2. Etched with Picric Acid. Troost-Sorbite. Fig. 104—Mag. 1000x. Tempered 7 days at 1158 Degrees Fahr. Scleroscope Reading = 25.8. Etched with Picric Acid. Ferrite + Carbide + Graphite. Fig. 105—Mag. 1000x. Tempered 7 days at 1192 Degrees Fahr. Scleroscope Reading = 24.5. Etched with Picric Acid. Ferrite + Carbide + Graphite. Fig. 106—Mag. 1000x. Tempered 7 days at 1192 Degrees Fahr. Scleroscope Reading = 24.5. Unetched, Showing Graphite. Fig. 107—Mag. 1000x. Tempered 7 days at 1192 Degrees Fahr. New specimen from hardened condition. Etched with Picric Acid. Ferrite + Carbide + Graphite. Fig. 108—Mag. 1000x. Drawn 6 days at 1238 Degrees Fahr. Another New Specimen from Hardened Condition. Scleroscope Reading = 27.5. Etched with Picric Acid. Ferrite + Carbide + Graphite.

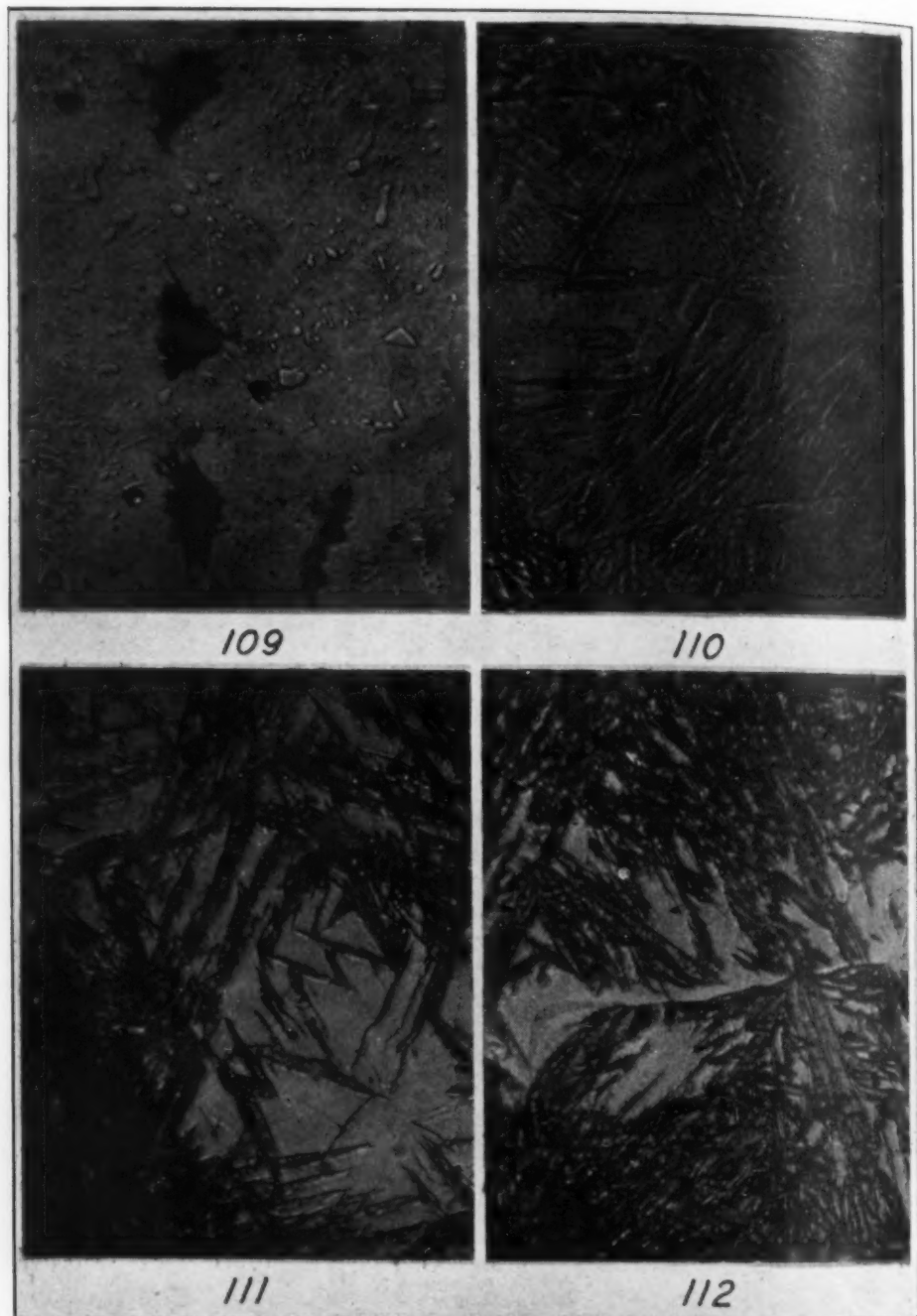


Fig. 109. Decomposition of Austenite on Tempering. Steels No. 6, High Carbon, 1.35 per cent Carbon. Quenching Treatment 2102 Degrees Fahr. 15 minutes, 5 per cent NaOH in Water at 25 Degrees Fahr. Fig. 109—Mag. 1000x. Tempered at 1240 Degrees Fahr. for 7 days. This Specimen was taken from Stock and Originally was a "Compromise Structure". Scleroscope Reading = 27.6. Original Scleroscope Reading = 42.1. Etched with Picric Acid. Ferrite + Carbide + Graphite. Figs. 110, 111 and 112—Decomposition of Austenite on Tempering. Steel No. 7. Quenching Treatment 2462 Degrees Fahr. 15 minutes, 5 per cent NaOH in Water at 25 Degrees Fahr. Specimens T. A. 1 and T. A. 2 were each Heated on Differential Curves with the Annealed No. 7 Steel as a Neutral Body. Fig. 110—Mag. 1000x. Center After Quench. Scleroscope Reading = 87. Etched with Picric Acid. Martensite + Austenite. Fig. 111—T. A. 1. Mag. 1000x. Heated in 1 Hour from Room Temperature to 815 Degrees Fahr. Scleroscope Reading = 74. Etched with Picric Acid. Troostite + Austenite. Fig. 112—T. A. 2. 1000x. Heated in 5 Hours from Room Temperature to 850 Degrees Fahr. Scleroscope Reading = 69. Etched with Picric Acid. Troostite + Austenite.

On tempering at 212 degrees Fahr. (100 degrees Cent.) there is a general increase in the specific gravity. Microscopic examination showed that during this period of tempering there was a tempering of the martensite needles. This leads to the conclusion that the increase in density is due to the predominance of the influence of the relieving of stress in the martensite, the growth of ferrite particles, and to the coalescence of carbide particles.

It will further be observed that the greatest increase in density is in steel No. 8 which was quenched in water, and the smallest increase in density is in steel No. 10 also water-quenched. Microscopic examination of these steels showed that the amount of martensite was in about the same proportion as the increase in specific gravity: the maximum amount in No. 8 and the minimum in No. 10. The microstructures further show that there is little or no change in the austenite at 212 degrees Fahr. (100 degrees Cent.). On tempering at 374 degrees Fahr. (190 degrees Cent.) there is a breakdown in the austenite which goes practically to completion in 4 days. Simultaneous with this breakdown of the austenite there is a decrease in specific gravity and a marked increase in hardness. The changes in microstructure lead to the conclusion that the increase in hardness and the decrease in density are due to the decomposition of austenite which results in a recrystallization of gamma to alpha, the formation of fine grains, and the formation of carbide particles in a finely divided condition. Tempering tests at still higher temperatures are known to result in decreases in hardness and increases in specific gravity. These changes as previously pointed out may be attributed to the increase in size of the alpha crystals, and the coalescence of the carbide particles.

*Results of Tempering Experiments*—The results of tempering tests indicate that:

a. In the carbon steels examined the martensite needles (light colored and in relief) showed progressive darkening at 212 degrees Fahr. (100 degrees Cent.) as the time at temperature increased. These tempering tests and the specific gravity tests indicate that precipitation of the carbide particles takes place slowly until the very dark troostite is formed. On continued tempering at higher temperatures, 392-518 degrees Fahr.

Carbon, 1.35  
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(200-270 degrees Cent.), there is a coarsening of the structure which is followed by a decrease in hardness probably caused by grain growth in the ferrite and coalescence of the carbide particles to a size that can be identified as a distinct phase.

b. The martensitic structures in alloy steels decompose on tempering similarly to carbon steels with a darkening of the needles but are more stable and either require a higher temperature for their decomposition or a longer time or both.

c. The decomposition temperature for an austenite is always higher than the decomposition temperature for the martensite in the same specimens or in steel of the same composition. The range of temperature of the austenitic decomposition for the steels included in this investigation are as follows:

Steel No.	Description	Range of Austenitic Decomposition Austenite to Troostite
5	High Nickel	625°C-670°C. (1158°F.-1240°F.) (Graphitizes)
1	Cobalt-chromium	525°C-562°C. ( 978°F.-1042°F.)
2	High Manganese	410°C-480°C. ( 770°F.- 896°F.)
3	Carbon-chromium	150°C-200°C. ( 302°F.- 392°F.)
4	High Speed	150°C-200°C. ( 302°F.- 392°F.)
6	High Carbon	150°C-200°C. ( 302°F.- 392°F.)
8	Tungsten Tool	150°C-200°C. ( 302°F.- 392°F.)
10	High Carbon	150°C-200°C. ( 302°F.- 392°F.)

d. It has been found possible to produce martensitic needles by tempering austenitic steels. These needles have been observed in the cobalt-chromium steel No. 1 and in the manganese steel No. 2. The exact conditions under which the martensite needles form on the tempering of these austenites has not been determined. However, it may be possible that there is an overlapping of the ranges of stability of the austenite and the martensite of steels of certain compositions. Frequently, dark needles (troostite) are produced in the tempering of austenitic steels.

e. Nodular troostite usually forms on tempering austenite. Its common places of formation are at grain boundaries (Fig. 66), along slip planes (Figs. 44, 45, 46, 81 and 88), or along martensite, cementite, or troostite patches (Figs. 74 and 75).



## COMPARATIVE TESTS ON BALL BEARING STEELS

BY T. L. ROBINSON

### Abstract

*This paper gives the results of a series of alternating stress tests in which an attempt was made to compare some of the alloy steels now used in ball bearing manufacture.*

*The alternating stress tests are supplemented by a static bending test designed to eliminate as far as possible any alignment factors.*

*The test results and subsequent microscopic examination indicated that the size and distribution of the particles of excess cementite were factors in the performance of the steel both under reversed stress or fatigue tests and under the static bending test. The steels in which these particles of cementite were comparatively small and uniformly distributed showed a markedly greater endurance and strength.*

FOR many years the steels used in the manufacture of high grade annular ball bearings have been standard in chemical composition, the chief alloy element being chromium. Recently this steel has been modified by several bearing manufacturers to include other elements such as molybdenum and vanadium. These changes were made in most cases only after extended service tests on bearings had indicated a superiority of these steels over the standard analysis. It was with the idea of accounting for these differences in performance from a metallurgical standpoint that "comparative tests on ball bearing steels", were undertaken.

The type analysis of the three steels considered are as follows:

Type of Steel	C	Cr	Mn	Sil	Phos & Sul	Mo	V
Chromium .....	1.00-1.10	1.30-1.50	0.20-0.40	0.15-0.30	0.025 max.	.....	.....
Chromium-molybdenum ..	1.00-1.10	1.10-1.30	0.20-0.40	0.15-0.30	0.025 max.	0.30-0.50	.....
Chromium-vanadium ...	1.00-1.10	1.30-1.50	0.20-0.40	0.15-0.30	0.025 max.	.....	0.20-0.40

The following designation will be used:

A—Chromium No. 1

B—Chromium No. 2

C—Chromium-vanadium No. 1

D—Chromium-vanadium No. 2

E—Chromium-molybdenum No. 1

F—Chromium-molybdenum No. 2

The author, T. L. Robinson, is metallurgist of the McGill Metal Company, Valparaiso, Ind.



Each type of steel was represented by the product of two different steel mills.

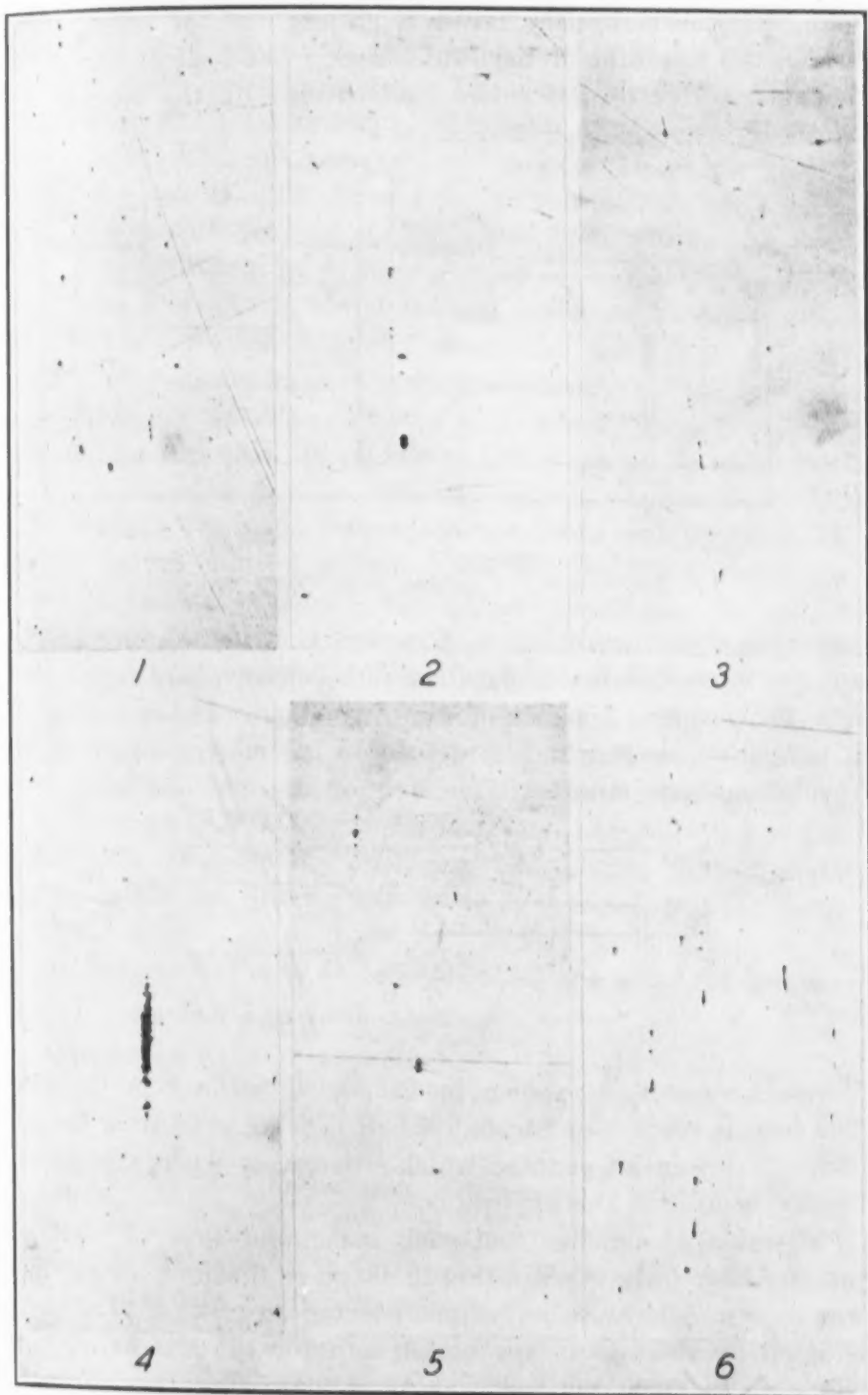
Every effort was made to eliminate the factor of nonhomogeneity due to blowholes, pipes, and nonmetallic inclusions. Bars were selected in which the inclusions were at a minimum as to size and number. This selection was made by cutting a specimen from each end of each bar and examining it under the microscope. Figs. 1 to 6 are a series of photomicrographs of polished unetched sections of these steels at 50 diameters magnification. The slag streak in steel D is exceptional, as the steel is in reality comparatively clean. Steel F is probably the dirtiest with little to choose from among the remaining five.

It was desirable to start with steels of similar hardness and microstructure; hence, the work was confined to hot-rolled, annealed round bar stock as supplied for the automatic screw machines. In these steels inspection of both hardness and microstructure gives an excellent control of the machining properties of the steel for which neither factor alone can be trusted. Variation of machinability in these steels, with the same tool setting, leads to high rejections for differences in dimension after turning. Hence it is specified that the bars have a Brinell number of from 175 to 200 and that the excess cementite be in a spheroidized condition.

The chromium-molybdenum and chromium-vanadium steels of this class were reported to have a greater toughness and a greater fatigue resistance than the standard chromium steel; hence, the endurance limits under reversed bending stresses were made the principal features for investigation.

Fig. 7 is a diagram of a machine built for this purpose. It follows closely the White-Souther machine in which a specimen is loaded and rotated as a cantilever beam. The advantages of this type of machine have been pointed out by Prof. Moore<sup>1</sup> and others and are too well known to dwell on except to point out that extreme simplicity is not the least of these advantages. This test is not entirely representative of conditions obtaining in service in ball bearing operation, because in the test there is a complete reversal of stress with every revolution. The maximum load stress on a bearing in operation is a repeated stress rather than a reversed stress. A fatigue effect would be most noticeable at the

<sup>1</sup>H. F. Moore and J. B. Kommers, University of Illinois, Bulletin No. 124, p. 24.



Figs. 1 to 6—Photomicrographs of Polished, Unetched Sections of Steels A, B, C, D, E and F Respectively at 50 Diameters Magnification.

point on the stationary member, either inner or outer, through which the resultant of the load passes. The load at this point is about one-fifth of the entire load carried by the bearing. The

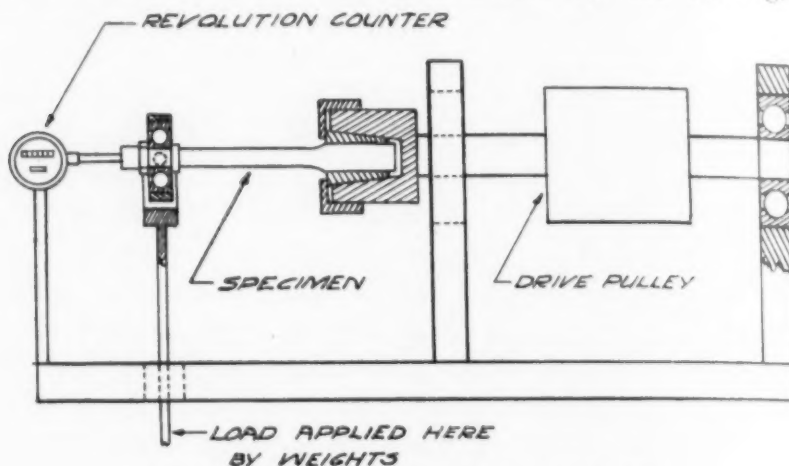


Fig. 7—Sketch of Rotating Beam Testing Machine.

reversed stress condition is, however, the more severe case and we are interested in comparative rather than quantitative values.

The machine consists of a spring chuck carried outboard on a horizontal shaft which is driven by an electric motor at 1400 revolutions per minute. One end of the test specimen of the

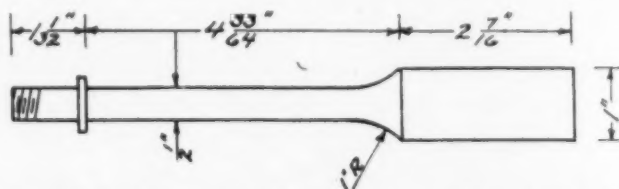


Fig. 8—Sketch of Specimen Used in Rotating Beam Testing Machine.

White-Souther type, shown in Fig. 8, is secured in the chuck. The load is suspended through a ball bearing yoke from the other end. A revolution-counter which disengages when the specimen breaks, completes the apparatus.

Twelve specimens from each make and type of steel were prepared for test. A variation in shape or finish of the specimens was known to have an important bearing on results. The method adopted in making up specimens both for the alternating stress test and the bending test was as follows: The specimens were turned in a lathe from a 2-inch round bar to  $0.5 \pm 0.040$  inch. All specimens were then given a similar and standard hardening

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treatment which consisted of quenching them in oil from 1510 degrees Fahr. and afterwards tempering in an oil bath at 350 degrees Fahr. for twenty minutes. The resulting Rockwell "C" scale hardness for all specimens fell within 60 plus or minus 1. It was believed that this variation in hardness would not greatly influence the results. The specimens were then finish-ground to nominal size within plus or minus 0.0002 inch with a No. 60-K wheel. This grinding operation eliminated any distortion due to hardening and also removed enough surface material to eliminate any decarburization. The specimens were then polished with 00 emery cloth to a uniformly high finish. A specimen was set up in the machine and run continuously to failure or to fifteen million reversals, which figure was accepted as an infinite run.

No attempt was made to complete a life curve to any distance either side of the point of critical stress, but the effort was rather to closely bracket this point. The critical stress may be defined as the highest stress at which a specimen would withstand an infinite number of reversals. This critical stress can be located with surprising accuracy and in almost all cases it represents the stress to which an addition of 2000 pounds would reduce the life to a few hundred thousand revolutions. In actual practice we ran the specimens to 15,000,000 revolutions only, but the marked flattening of the curve at the critical stress enabled us to predict by interpolation an infinite life for a specimen operated below this stress.

The endurance limits thus established are given as follows:

ENDURANCE LIMITS		
Designation	Type of Steel	Stress in lbs. per sq. in.
D	Chromium-vanadium	95,000
C	Chromium-vanadium	94,000
B	Chromium	88,000
A	Chromium	84,000
F	Chromium-molybdenum	83,000
E	Chromium-molybdenum	75,000

The wide variation in these results on steels of different heats but similar chemical compositions, is at once apparent.

The fractured surface of the specimen was examined in all cases. There was, however, little apparent difference as all showed the silky sheen characteristic of well-hardened ball bearing steel with the possible exception of E. This specimen showed slight



traces of segregation in the nature of fine fracture planes which would be barely discernible to an experienced eye.

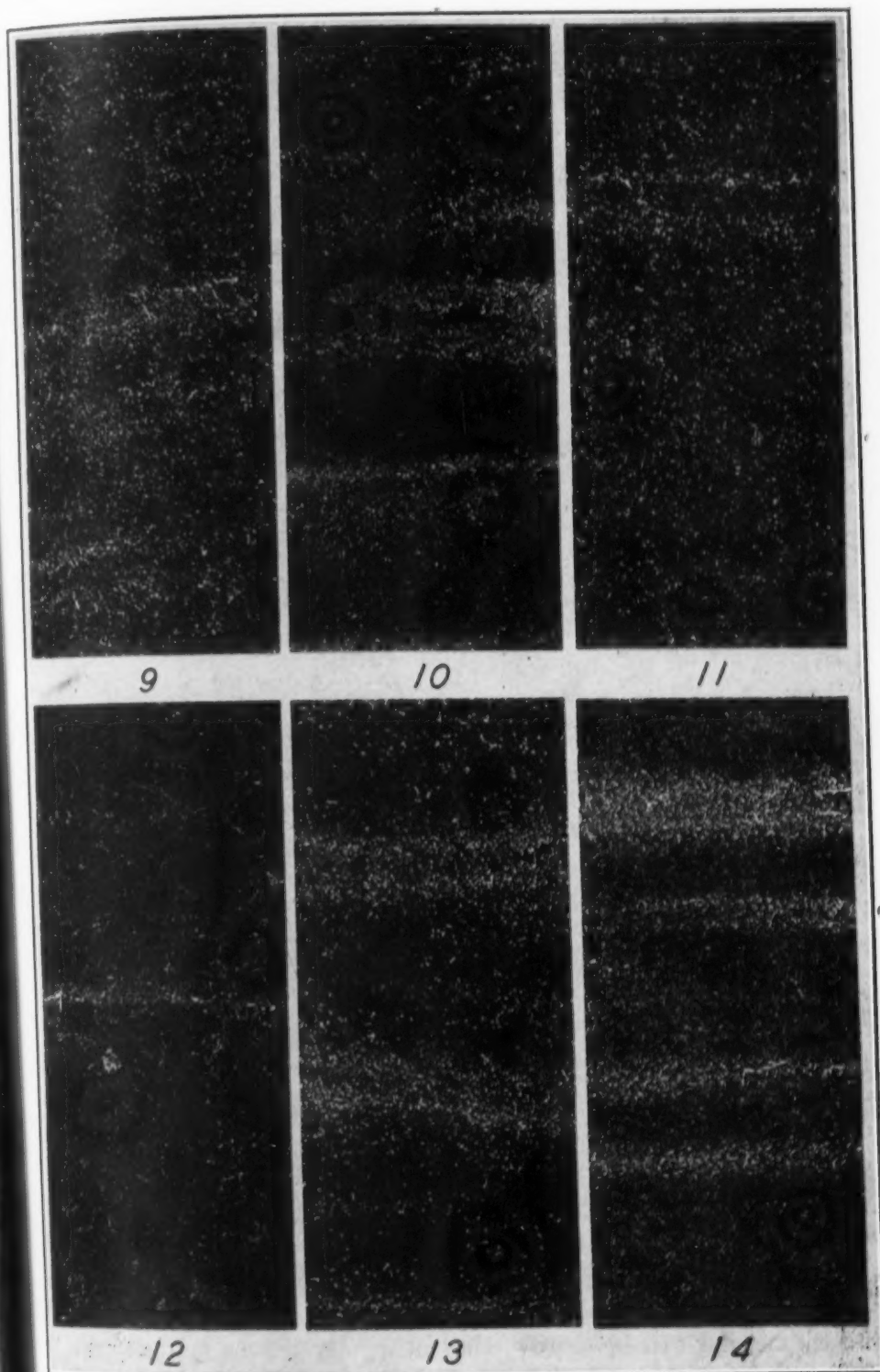
The microstructure of the specimens was next examined which investigation revealed the probable cause for the variation in performance. There was a tendency for the steels which showed a marked segregation of the carbides to give low results. This effect is apparently independent of chemical composition except as it affects segregation in the steel. The mechanism of segregation in high speed steel has been pointed out by Grossmann and Bain<sup>2</sup> and their analysis undoubtedly applies to the steel under consideration to some extent.

This structure is probably the result of the difference in carbide concentration in the original structure of the ingot. The carbide-rich areas so formed become elongated in rolling and exist in the bar stock as roughly cylindrical streaks lying parallel to the axis of the bar. This condition is usually much more marked near the center of the bar than toward the outside. The only effective cure for this segregated condition is a considerable amount of hot work or deformation. The appearance of the six steels under consideration is shown in the photomicrographs, Figs. 9 to 14 at 100 diameters magnification and in Figs. 16 to 21 at 500 diameters magnification. The series of photomicrographs is introduced for the purpose of emphasizing the distribution and size of the carbide globules. The martensitic ground mass is structureless at the magnifications and etching used.

This steel is in the hardened condition as used in ball bearings in service. The chief constituent is martensite, although there is some austenite and probably some troostite present. The grain size of the martensite being sub-microscopic, we attribute any network appearance to carbides at the former gamma iron grain boundaries. The carbide particles represent an excess which cannot be held in solution in the alpha iron. The visible carbide particles are of two sizes. The larger and more irregularly shaped are carbides which have persisted through the cogging and rolling operations from the dendritic ingot structure. The smaller round particles are carbides which have been precipitated in cooling from some previous processing. There is probably in addition a sub-microscopic precipitation of carbide to which part of the hardness of the material is due.

<sup>2</sup>Marcus A. Grossmann and Edgar C. Bain, *Journal, Iron and Steel Institute*, 1924.





Figs. 9 to 14—Photomicrographs of Polished and Etched Longitudinal Sections of Steels A, B, C, D, E and F Respectively at 100 Diameters Magnification Deeply Etched in Alcoholic Nitric Acid to Develop the Cementite.

It is quite generally recognized that a steel in which the carbide is evenly distributed throughout the martensite would be more desirable than one in which the carbide was segregated in bands. It is also theoretically desirable that the carbide particles be small and evenly distributed rather than large and evenly distributed. The former certainly represents a more nearly homogeneous condition and would allow of easier solution of the carbides at the hardening temperatures. It is possible that resistance to wear in service is improved by a uniform distribution of small carbide particles. Any wearing action would tend to erode the softer matrix leaving the carbide particles in relief. The rate of wear would then be determined by that of this harder constituent and a uniformly slower rate be expected from an even distribution of that constituent.

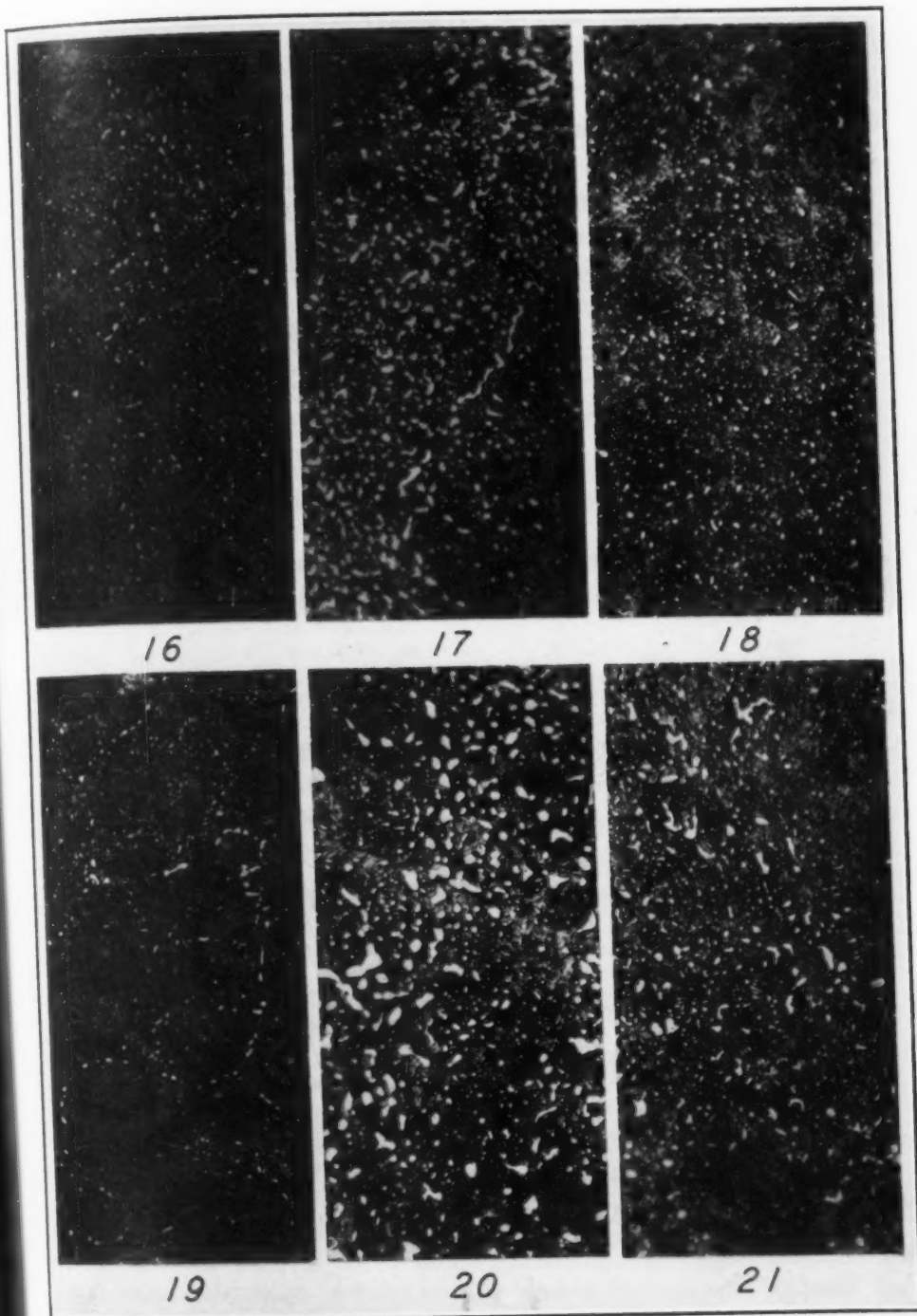
Messrs. Desch and Roberts<sup>3</sup> reach a somewhat similar conclusion with respect to the effect of cementite distribution on resistance to wear in high carbon chromium razor steels.

Figs. 9 to 14 will give an idea of the general distribution of the segregation, and Figs. 16 to 21 of the size of the particles going to make up this banded appearance. It is quite apparent in considering photomicrographs in Figs. 9 to 14 that steels E, F and B are conspicuously banded. The prominence of the banding of the specimens stand about in the order above mentioned, the first referred to being the worst. There is practically no sign of banding in specimens A, C and D. While the nature of the evidence does not lend itself to a precise rating of the specimens as to degree of segregation, the extremes can be picked with certainty. Thus A and C show no traces of segregation and E is certainly the worst of the group in this respect.

In Figs. 16 to 21 it will be noticed that in steels E, F and B, the carbide particles are large, irregular in shape and unevenly distributed through the martensitic ground mass. The particles in A, C and D are relatively small and evenly distributed.

Since the steel E gave the lowest values in both the fatigue and bending test and since D gave high values in both tests and since the physical test results on the other steels dovetail nicely into positions interpolated for them from their microstructures the conclusion can be drawn, that other conditions being constant,

<sup>3</sup>Cecil H. Desch and Aubrey T. Roberts, *Journal, Iron and Steel Institute*, 1923, No. 1, p. 258.



Figs. 16 to 21—Photomicrographs of Polished and Etched Longitudinal Sections of Steels A, B, C, D, E and F Respectively at 500 Diameters Magnification Deeply Etched in Alcoholic Nitric Acid to Develop the Cementite.

resistance to fatigue in this type of steel is considerably lowered by pronounced carbide segregation.

These data are confirmatory of some of the previous work on endurance of steel under alternating stresses. Gillett and Mach<sup>4</sup>

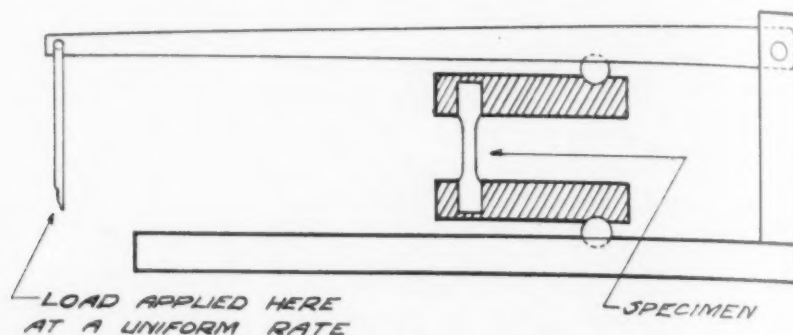


Fig. 22—Diagrammatic Sketch of Bend Testing Apparatus.

state in summarizing, that claims for favorable influence of molybdenum and vanadium on endurance appear to have been exaggerated.

Practically all of the investigators agree that dirty steel has a lower resistance to fatigue or at least is erratic in behavior under alternating stresses. While these steels are exceptionally clean as regards slag and inclusions and though the carbide can-

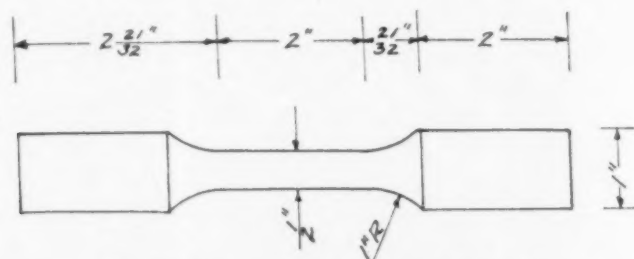


Fig. 23—Sketch of Specimen Used in Bend Tests.

not be classed as an inclusion, its effect on the endurance performance may well be similar to that of a nonmetallic inclusion. In this way the results can be reconciled with the above mentioned observations.

Prof. Moore<sup>5</sup> and other investigators show that the static tests as carried out on the ordinary tension compression testing

<sup>4</sup>H. W. Gillett and E. L. Mach, *Proceedings of American Society Testing Material*, Vol. 24, Part II, p. 544.

<sup>5</sup>H. F. Moore and J. B. Kommers, *University of Illinois, Bulletin No. 124*, p. 95.



machine, are not closely indicative of the performance of the steel under alternate stresses. The ultimate strength is apparently of slightly more value than the elastic limit in predicting fatigue limit, possibly since the errors of determination are less. Since no close correlation of values was expected and with the idea of eliminating as many variable factors as possible a very simple bending test was arranged. The specimens shown in Fig. 23 were fixed at both ends in special holders, to which a load was applied by weighting at a uniform rate a lever acting through ball swivel joints. This scheme is shown diagrammatically on Fig. 22. Unknown factors such as misalignment are eliminated and the loads can be definitely computed. When direct comparisons rather than quantitative results are desired, this arrangement has much to recommend it. Results are as follows:

Designation	Type of Steel	Unit Stress in lbs. per sq. in.	Rockwell "C" Scale Hardness
C	Chromium-vanadium	290,000	59.5
D	Chromium-vanadium	285,000	60.5
A	Chromium	233,000	60.5
B	Chromium	220,000	59.5
F	Chromium-molybdenum	218,000	61.5
E	Chromium-molybdenum	164,000	61.5

There is no apparent agreement in values between the two types of tests but the resulting positions of each steel in both tests are practically the same as shown in the following table:

#### *Fatigue Test*

D—Chromium-vanadium	A—Chromium
C—Chromium-vanadium	F—Chromium-molybdenum
B—Chromium	E—Chromium-molybdenum

#### *Bending Test*

C—Chromium-vanadium	B—Chromium
D—Chromium-vanadium	F—Chromium-molybdenum
A—Chromium	E—Chromium-molybdenum

These test results offer some confirmation of actual service tests and experience with bearings in operation. On several instances bearing rings have been put in service, made from steel showing heavy segregation. The results pointed unmistakably to the inferiority of this material.

It is also interesting to note that of a comparatively small



percentage of bearings returned from service as defective in material, the majority of those returned show this excessive carbide segregation. Some manufacturers have for some time considered the elimination of this structure sufficiently important to put in effect a bar-to-bar microscopic inspection.

The results obtained apply more specifically to the members of the bearing which receive little or no hot or cold work during fabrication. Thus most of the smaller race rings, say up to 2.5 inches in diameter, are turned in automatic screw machines and receive no working or deformation. The so-called "turned" balls (i. e., balls which are cut from spherical form bar stock) fall under the same category.

Rings varying in size from 2.5 inches to about 4 or 5 inches are in many instances hot upset from bar stock. The same applies to probably a majority of the balls now manufactured in the U. S., i. e., they are hot or cold upset from bar stock or wire. In these cases the actual amount of the reduction or deformation determines whether the structure of the bar stock is broken up and to what extent the test results apply. Many instances of upset forged rings in which the segregation was displaced rather than broken up have been seen. The rings above 5 inches are in most cases hammer forgings. The amount of hot work in these cases is always sufficient to break up segregation in this bar stock. Hence the conclusion to be drawn both from laboratory and service tests is that the quality of the steel is largely determined by the size and distribution of the cementite.

The author wishes to thank F. R. Schubert formerly of the Strom Ball Bearing Manufacturing Company, Chicago, for readily granting the facilities for this investigation. He also is indebted to W. P. Sykes of the National Lamp Works, Cleveland Ohio, for the photomicrographs at 500 diameters and to O. V. Cederberg of the McGill Metal Company for supervision of most of the tests.

RELATION OF SURFACE TO VOLUME IN CRYSTALS AS  
A DETERMINING FACTOR IN GRAIN  
GROWTH OF METALS

BY G. W. WALKER

*Abstract*

*In this paper an attempt is made to obtain a more critical view of the mechanism involved in the grain growth of metals taking into consideration the physical factors; surface, volume, surface tension, surface energy, solubility, boundary migration, etc. Also an attempt is made to consider these phenomena in a quantitative aspect as far as data is available. In connection with this discussion some of the statements of Jeffries and Archer as given in their book, "The Science of Metals", are considered and criticized.*

THE relation of surface to volume as an important and determining factor in affecting the physical properties of crystalline aggregates has been largely overlooked by metallurgists. In discussing the "Possible Causes of Grain Growth," Jeffries and Archer in "Science of Metals", 1924, p. 119, state: "The atoms at the surface of a liquid or solid body are less subject to the attraction of their neighbors than are the atoms in the interior. They are, therefore, more free to move and, consequently, possess more energy of motion. A 'free' surface of any body is thus a locus of extra energy, known as 'surface energy' or sometimes 'surface tension.'" Now exactly what is meant by the statement that "the atoms at the surface of a liquid or solid body are less subject to the attraction of their neighbors than are the atoms in the interior"? Is molecular attraction a variable force, as the above statement implies, or is it constant? Any given atom in the interior of a liquid has a certain number of neighbors within the limits of molecular attraction, under given conditions of temperature and pressure. If this atom is now moved to the surface of the liquid it will have only half as many neighbors within the radius of molecular attraction. This attraction is now exerted along the surface by neighbors in the surface and inward in all directions by neighbors below the surface. The

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effect at the surface is that of "an elastic membrane covering the surface." The pull of this elastic membrane gives rise to surface tension. Hence the correct statement of the facts is not that the atoms at the surface are "less subject to the attraction of their neighbors," but that they are subject to the attraction of fewer neighbors and are thus more free to move. The statement that: "They are therefore more free to move and, consequently, possess more energy of motion," implies that energy of motion is a result of being more free to move. In what way does a body more free to move possess more energy of motion, and if it does possess more energy, how does it get it?

The surface tension of a liquid decreases as the temperature rises and becomes zero at the critical temperature of the liquid. At the critical temperature of the liquid, a molecule of the liquid is most free to move. As the temperature falls, surface tension increases, and a molecule of liquid at the surface is less free to move, as it now has more neighbors within the molecular radius of attraction. In order to give the molecule greater freedom of movement, heat must be added to the liquid. In other words, freedom of movement results from the addition of energy in the form of heat. Therefore, more energy of motion results from the addition of more energy in the form of heat, and as a result the "atoms are more free to move."

Consequently, the "atoms" are more free to move because they have received more energy of motion and not as stated: "They are therefore more free to move and, consequently, possess more energy of motion."

The statement is made that: "A 'free' surface of any body is thus a locus of extra energy, known as 'surface energy' or sometimes 'surface tension.' " Surface tension and surface energy are not the same thing as is implied in the above statement. As stated by Willows and Hatschek, "Surface Tension and Surface Energy," p. 6, ". . . it is not correct to define the surface tension as surface energy per square centimeter."

Work is required to produce an increase in surface. If the temperature remains the same, the work goes to increase the surface energy. In addition, heat must flow into the surface film to keep the temperature constant. This also increases the energy. "The total energy of the surface, therefore, consists of two terms,

of which one represents the amount of work done against surface tension, and the other the inflow of heat during the extension of the surface."

On page 119, "Science of Metals," the authors conclude from the statements previously quoted and examined, that: "There is, therefore, a tendency for small particles, such as a drop of a liquid or crystals of a solid, to unite to form larger particles, since in doing so they diminish their surface area and hence their total energy."

It is difficult to see how the above conclusions can be clearly shown to follow from the previous statements. It is nothing more than a restatement of the opening sentence of the paragraph, that: "Grain growth obviously results from the tendency of matter to assume the form of greatest physical stability, which is the form of least energy." This reminds one of Herbert Spencer's definition of evolution as the integration of matter with the concomitant dissipation of energy.

On page 121 et seq., the authors say: "The factors affecting the stability of grains of a metal, and hence the growth force, have been indicated (size, shape, surface, and distortion)." Regarding "Shape and Surface of Grains," they say: "There are frequently found, even in unstrained metals, grains of such elongated or extended shape that the ratio of surface area to volume is abnormally high for grains of that size. Presumably, such grains are less stable than grains of the same size but of more nearly equiaxed shape, which, therefore, tend to grow at the expense of the former." What the authors mean to say is that of all bodies of equal volume, the sphere has the least surface in proportion to the volume, that is, the ratio of surface to volume is the lowest. Of all grains that have the same volume, the grain that departs the most widely from the spherical shape will have the highest ratio, that is, the greatest extent of surface in proportion to the given volume. Such grains, owing to greater extent of surface in proportion to volume, will be less stable than grains of equal volume having smaller surfaces, and hence the latter will "tend to grow at the expense of the former."

The reason for this tendency is stated as follows: "The tendency of metals to reduce their store of energy by increasing their grain size is the factor which has probably been considered the



principal cause of grain growth." Also, "The authors believe that grain size . . . furnishes the basis for a completely satisfactory theory of grain growth."

Jeffries and Archer have not fully and clearly developed the implications contained in their statements. If grain growth takes place by boundary migration, as appears fairly conclusively shown, then the relations of surface, volume, surface energy, and surface tension are of fundamental importance in reaching any adequate explanation of grain growth.

Is decrease in surface area and hence total energy a sufficiently clear explanation of the tendency of smaller particles to unite to form larger particles? If this tendency as a universal law exists, then the matter of the universe must finally be collected into one mass—a perfect sphere. On the other hand, the smaller the particle, the greater the surface area and the greater the total energy. Hence an infinitely small particle would have an infinitely large surface in proportion to volume and an infinitely large total energy. Either way the explanation does not seem entirely adequate.

On the basis that metals generally crystallize in the form of cubes, the following results have been calculated as to the relation of total surface to volume as the crystals decrease in size:

Edge of Cube	Ratio of Surface to Volume
1.000"	6:1
0.125	48:1
0.0075625	768:1
0.0009453	6104:1
0.000059	97664:1

Here we have shown the interesting and surprising fact that in grains of microscopic size, the ratio of surface to volume must be relatively enormous.

Along with the great increase in surface there must correspondingly go great changes in surface phenomena, that is, changes in surface tension, surface energy, solubility, boundary migration, etc.

The very large extension of surface in grains of small size, render them much more open to attack and disintegration as compared with grains of larger size. Hence in solutions small grains are more rapidly dissolved than large ones. In the case of the inflow of heat, the larger surface in proportion to volume allows of more rapid absorption of heat and hence more rapid



solution and disintegration as the temperature is raised.

An important surface effect is the relation between melting point and size of particles. Pawlow, as quoted by Willows and Hatschek, found that "granules of salol with a surface of  $230\text{--}1300\mu^2$  have a melting point  $2.0$  degrees Cent. lower than particles with a surface  $100$  times greater, i. e., with ten times greater diameter."

As stated by Willows and Hatschek, "Surface Tension and Surface Energy," p. 23: "osmotic pressures are proportional to the number of molecules dissolved in the same volume, or in other words, to the solubilities of large and small particles respectively, and will be different if these solubilities are different."

In the case of calcium sulphate it was found by Hulett, that the

"solubility . . . . . was  $18.2$  millimoles per liter for particles of a radius,  $a = 0.00003$  cm., and  $15.33$  millimoles for particles of a radius,  $a = 0.0002$  cm., so that the smaller particle shows considerably larger solubility."

Since osmotic pressures are proportional to the amounts dissolved, the smaller particles show higher osmotic pressures. As Willows and Hatschek show: "since the big drop has a smaller surface than equal volume of small ones," the transfer of a quantity of a liquid from one to the other "leads to a decrease of surface energy and at the same time liquid is taken from a place of higher to a place of lower vapor pressure. The growth of particles in a precipitate or in a supersaturated solution is found to occur in a similar manner, i. e., large crystals grow at the expense of small ones; but there we have to deal with the decrease in energy of a surface solid-liquid . . . . and a transfer of matter from a place where the osmotic pressure is higher to one where it is lower."

Decrease in size of crystals is no doubt accompanied by changes in surface tension as the ratio of surface to volume increases. Also, as Willows and Hatschek say: "we know that the surface tension of a liquid increases with falling temperature, and it is, therefore, improbable that it should suddenly disappear when the temperature falls to the freezing point and the liquid changes to a solid." The change from the liquid to the solid state and decrease in grain size should bring about high surface tension. Surface tension is accompanied by internal or

intrinsic pressure. According to Willows and Hatschek: "... high intrinsic pressures must be accompanied by high surface tensions, since surface tension is a manifestation of the same cohesive force as causes intrinsic pressure. High surface tension should be accompanied by low compressibilities." Water which has low compressibility is shown to have an intrinsic pressure of about 11,000 atmospheres.

The surface tension of calcium sulphate has been found to be 1100 dyne per centimeter and of barium sulphate 4000 dyne per centimeter. As Willows and Hatschek say: "these figures entirely confirm the conclusion to which we have come on general grounds, that the surface tension of solids must have high values."

Reduction in grain size in metals is brought about by heat or heat and work. In any case grain refinement is brought about by the use of energy, thermal or mechanical or both. The grain refinement obtained is the equivalent of the amount of energy necessary to bring about the increase of surface to volume necessary for the refinement.

This energy may be considered to be stored in the crystals obtained in the form of surface tension, surface energy, osmotic pressure, and intrinsic pressure. Small crystals are more unstable than large ones in proportion to the amount of energy required to bring about the refinement. Thus we arrive at a more critical and fundamental insight as to the causes and conditions of grain growth.

#### DISCUSSION

**Written Discussion:** By Dr. Zay Jeffries and R. S. Archer, Cleveland.  
*From a letter addressed to the author of the paper.*

"We have read with much interest your manuscript 'The Relation of Surface to Volume in Crystals as a Determining Factor in Grain Growth of Metals.' The ideas which you have expressed seem to us to be substantially the same as our own as given in 'The Science of Metals.' We believe that you will find that the chief difference is in the manner of expression and in the fact that you have considerably expanded the idea of relation of surface to volume and have given more thought to quantitative considerations.

"On page 619 you question the following sentence: 'The atoms at the surface of a liquid or solid body are less subject to the attraction of their neighbors than are the atoms in the interior.' This statement is not meant to imply a variable force as you have interpreted it. Its meaning will perhaps be more clear if for the words 'subject to' we substitute the expression 'dominated by.'

"Referring to the use of the terms 'surface tension' and 'surface

energy' it is apparent that there is a distinction since one represents a force and the other energy. It is a fact, however, as stated in the book, that these terms have often been used interchangeably. The actual statement in the book is as follows: 'A *free* surface of any body is thus a locus of extra energy, known as *surface energy* or sometimes *surface tension*.' It is also a fact that neither the surface energy nor the surface tension is actually known for any solid substance. The whole subject therefore as regards solids comes, at the present time, under the head of qualitative phenomena.

"On page 622 you state: 'In the case of the inflow of heat, the larger surface in proportion to volume allows of more rapid absorption of heat and hence more rapid solution and disintegration as the temperature is raised.' This is not in accordance with our ideas; in fact it seems to us that a coarse grained metal should be a better conductor of heat than a fine-grained metal on account of the greater resistance to conduction at the grain boundaries.

"You have referred to the effect of particle size on the solubility of calcium sulphate. An example of this is also given in 'The Science of Metals' on page 122. Apparently our example is the same as the one you have quoted, although the figures for solubility do not check exactly when transposed from one unit to the other.

"You have also given an example of the effect of particle size on melting point referring to Salol. You will note on page 124 in 'The Science of Metals' an example of the effect of particle size on the melting point of myristic acid.

"You have evidently given considerable thought to this subject and have made a very clear statement. It is to be hoped that your contribution will stimulate thought by others and perhaps result in valuable changes or additions to the theory of grain growth."

## Educational Section

These Articles Have Been Selected Primarily For Their Educational  
And Informational Character As Distinguished From  
Reports Of Investigations And Research

### THE CONSTITUTION OF STEEL AND CAST IRON PART VIII

By F. T. SISCO

#### *Abstract*

*The present installment, the eighth of the series, discusses the constitution of the iron-carbon alloys containing 4.30 to 6.67 per cent carbon. These alloys are the hypereutectic cast irons and are made up of an eutectic of saturated austenite and cementite and the excess constituent, cementite. The constitutional changes in this series of alloys when cooled from the molten state to atmospheric temperature are described. In conclusion, the behavior of the whole series of iron-carbon alloys containing 0.01 to 6.67 per cent carbon, in cooling from a high temperature, is reviewed briefly.*

IN the previous chapter we studied the constitution of the cast irons containing between 1.70 per cent and 4.30 per cent carbon, equivalent to 25.5 to 64.5 per cent cementite. When solidification is complete these alloys are made up of two constituents, saturated austenite or gamma iron containing 25.5 per cent iron carbide in solid solution and cementite (iron carbide). We saw that saturated austenite and cementite form an eutectic or lowest melting alloy containing 4.30 per cent carbon and consisting of 48 per cent saturated austenite and 52 per cent cementite. If the alloy contains less than 4.30 per cent carbon it will contain less than 52 per cent cementite, hence there will be some saturated austenite in excess of the eutectoid ratio and we will have a hypoeutectic alloy. Instead of solidifying at a constant temperature this alloy will have a freezing range or mushy stage. Freezing will begin by the separation of the excess saturated austenite.

As soon as the hypoeutectic cast irons are completely frozen,

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transformations in the solid state take place. The first transformation is the rejection of cementite from the saturated austenite as the alloy cools from 2065 degrees Fahr. (1130 degrees Cent.) to 1300 degrees Fahr. (700 degrees Cent.), until when this latter temperature is passed the rejection is complete and the saturated austenite has been transformed into an aggregate of six parts of pearlite and one part of pro-eutectoid cementite.

#### SOLIDIFICATION AND COOLING OF EUTECTIC CAST IRON

Referring to Fig. 44 which has been reproduced from the last installment, it will be noted that the point B represents the composition of the eutectic iron-carbon alloy. This alloy contains 4.30 per cent carbon equivalent to 64.5 per cent cementite. We studied the characteristics and structure of this alloy in the previous installment.

To trace the constitutional changes taking place in the eutectic cast iron we will take the ordinate N, Fig. 44, and will follow the cooling of the alloy represented by this vertical line. At any temperature above B, 2730 degrees Fahr. (1500 degrees Cent.) for example, the alloy is a molten solution of carbon (or iron carbide) in iron. This solution cools unchanged until a temperature of 2065 degrees Fahr. (1130 degrees Cent.) is reached, point B, Fig. 44. This is the freezing point of the eutectic; the whole mass of metal freezes unselectively at this temperature. Immediately after solidification the solid eutectic consists of 48 per cent saturated austenite and 52 per cent eutectic cementite.

As the alloy cools from 2065 degrees Fahr. to 1300 degrees Fahr. (700 degrees Cent.) the austenite expels the proeutectoid cementite as it cools along the line ES (Fig. 44) until at 1300 degrees Fahr. the remaining austenite now of eutectoid composition is transformed into pearlite. When all of the transformations have taken place a cast iron containing 4.3 per cent carbon will contain 41 per cent pearlite, 6.6 per cent proeutectoid (excess) cementite, and 52 per cent eutectic cementite. The ultimate composition will be 35.5 per cent ferrite and 64.5 per cent cementite.

#### SOLIDIFICATION AND COOLING OF HYPEREUTECTIC CAST IRON

As representative of the hypereutectic cast iron series we will trace the solidification and constitutional changes in cooling



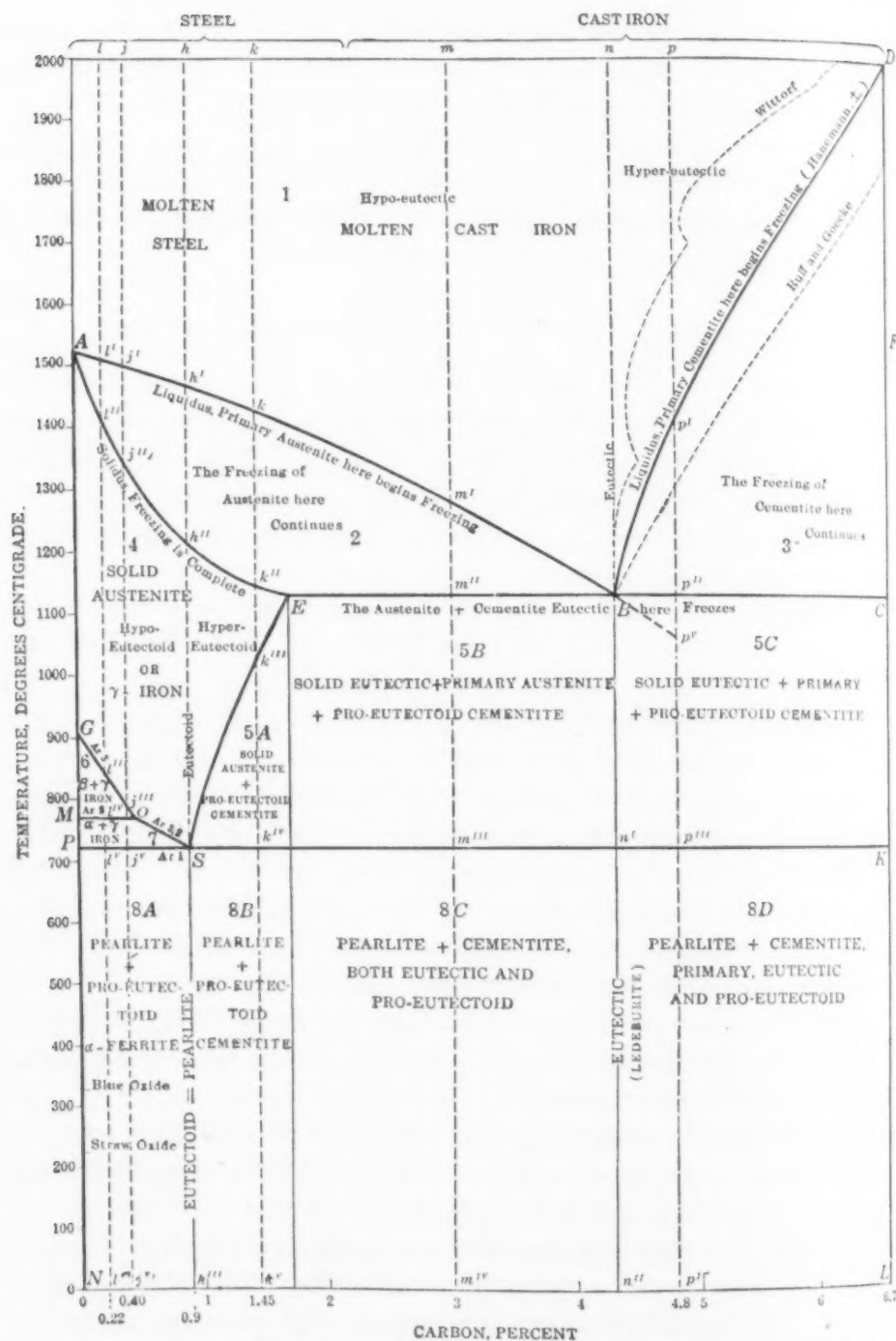


Fig. 44—The Metastable or Cementite-Austenite Iron-Carbon Diagram (Howe).

of an iron-carbon alloy containing 4.80 per cent carbon, ordinate P, Fig. 44. At a temperature of 2910 degrees Fahr. (1600 degrees Cent.) the alloy is a molten solution of 4.8 per cent carbon

or 72 per cent iron carbide in iron. From this temperature the alloy cools unchanged and at a uniform rate until 2585 degrees Fahr. (1420 degrees Cent.) is reached. Here an arrest in the cooling occurs. At this point the cementite in excess of the eutectic ratio starts to solidify. This excess cementite continues to precipitate out of solution as the alloy cools from 2585 degrees Fahr. to 2065 degrees Fahr., following the line DB, Fig. 44.

The precipitation of the excess cementite impoverishes the remaining melt in carbon until it has reached the eutectic percentage, when the whole mass solidifies. At the point B the composition of the cast iron is 79 per cent eutectic and 21 per cent of primary (excess) cementite. As the alloy cools from 2065 degrees Fahr. (1130 degrees Cent.) to 1300 degrees Fahr. (700 degrees Cent.) the austenite rejects the proeutectoid cementite until at 1300 degrees Fahr. it has reached the eutectoid composition and is transformed into pearlite.

#### CONSTITUTION OF HYPEREUTECTIC CAST IRON

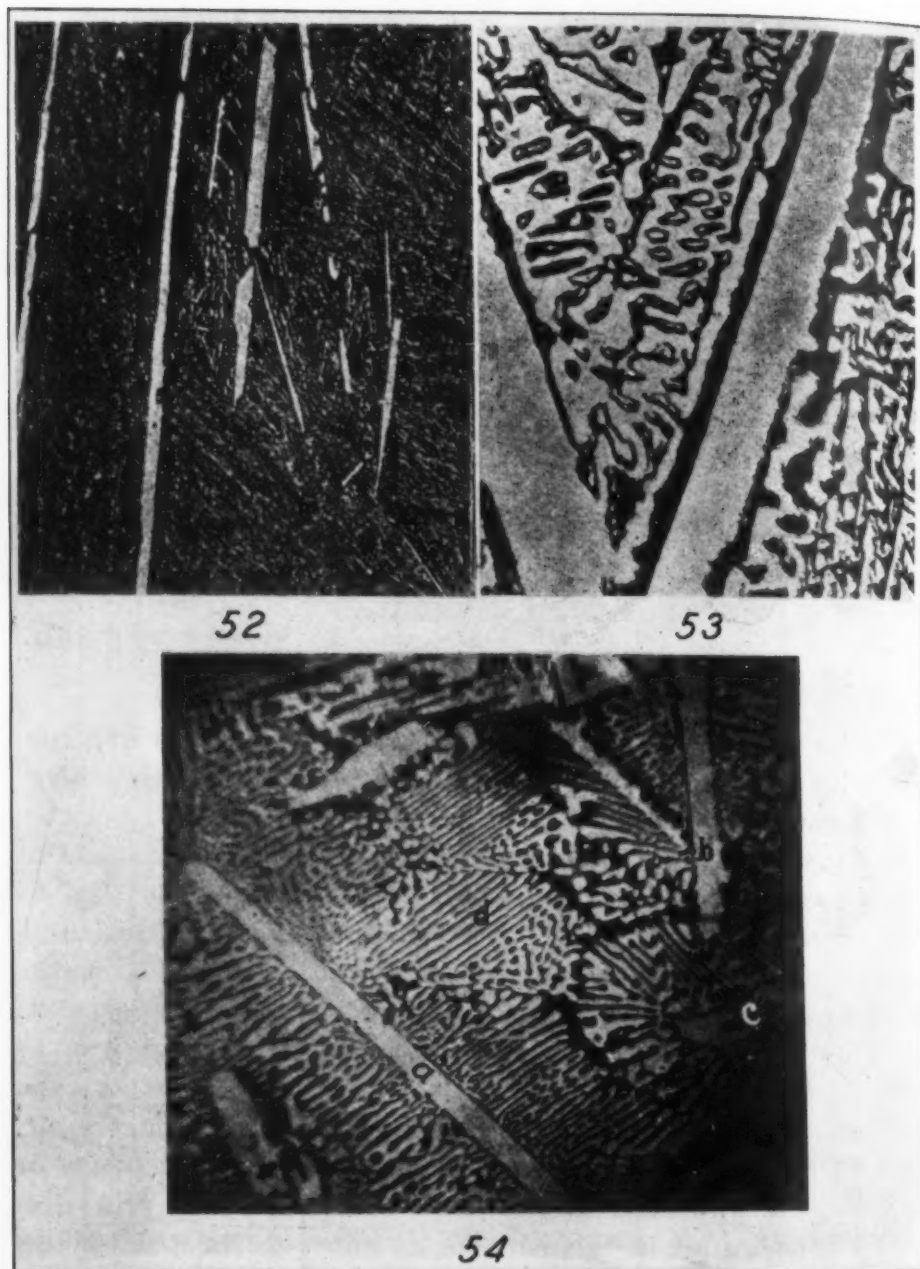
Immediately after solidification a hypereutectic cast iron containing 4.8 per cent carbon consists of 38 per cent eutectic austenite, 41 per cent eutectic cementite, and 21 per cent primary cementite. The structure of such an iron is shown in Fig. 52 and at higher magnification in Fig. 53 from Sauveur<sup>87</sup> and in Fig. 54 from Howe.<sup>88</sup> The structure consists of a matrix of austenite-cementite eutectic and islands of free cementite. The free cementite ordinarily occurs as hard white needles embedded in the eutectic.

If we compare the structure of hypereutectic cast iron as shown in Figs. 52, 53 and 54 with the structure of hypoeutectic cast iron of Figs. 48 and 49<sup>89</sup> we will see some striking similarities. In both cases the matrix is made up of parallel plates or honeycomb of the austenite-cementite eutectic. The free constituent is embedded in this eutectic as spines or isolated islands. In our hypereutectic alloy the excess constituent (cementite) takes the form of needles; in the hypoeutectic alloys the excess constituent (saturated austenite) crystallizes as elongated islands

<sup>87</sup>Metallography and Heat Treatment of Iron and Steel, 1926, p. 432.

<sup>88</sup>Metallography of Steel and Cast Iron, 1916, plate 1.

<sup>89</sup>Installment VII, this series, TRANSACTIONS, American Society for Steel Treating, February, 1927, page 290.



Figs. 52 and 53—Iron-Carbon Alloy. Hypereutectic. Structure Immediately After Solidification. Needles of Cementite in a Matrix of Austenite-Cementite Eutectic. (Sauveur.) Mag. 65 and 980X respectively. Fig. 54—Hypereutectic Cast Iron. White Bands a, b, c of Primary Cementite; Honeycomb of Eutectic. 30X. Howe.

many of which have a pine-tree structure. The hypereutectic cast irons may be readily distinguished from the hypoeutectic alloys by the appearance and shape of the excess constituent:

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<sup>30</sup>The  
<sup>31</sup>See  
page 292.

excess cementite occurs as hard, white needles; excess austenite etches dark and is more irregular in shape.

Our hypereutectic cast iron containing 4.8 per cent carbon immediately after solidification contains 79 per cent eutectic and 21 per cent primary (excess) cementite. The eutectic which makes up 79 per cent of the structure is composed of 38 per cent eutectic austenite and 41 per cent eutectic cementite. The cementite, both excess and in the eutectic, cools unchanged<sup>90</sup> to atmospheric temperature. The austenite undergoes transformation during the cooling, the cementite in excess of the eutectoid ratio being rejected by the solid solution.

After all transformations are complete our hypereutectic alloy will have the following structural composition.<sup>91</sup>

	Per Cent
Pearlite	32
Proeutectoid Cementite	5.2
Eutectic Cementite	41
Primary Cementite	21

From this we see that the hypereutectic cast irons will have four classes of cementite present: (1) the primary cementite which was in excess of the eutectic ratio; (2) the cementite in the eutectic; (3) the cementite rejected from austenite and called proeutectoid cementite; and (4) the cementite in the pearlite aggregate.

#### SUMMARY OF CONSTITUTIONAL CHANGES IN THE IRON-CARBON ALLOYS

In order to view clearly the constitutional changes taking place in the whole series of iron-carbon alloys we will summarize briefly, tracing the cooling of pure iron and seven alloys, represented on Fig. 44 by the ordinates I, J, H, K, M, N and P.

1—At 3000 degrees Fahr. pure iron is molten. It cools uniformly until 2785 degrees Fahr. is reached (point A, Fig. 44) at which constant temperature the whole melt solidifies as gamma iron. From this point cooling is uniform and uninterrupted until a temperature of 1650 degrees Fahr. is reached (point G, Fig. 44) at which temperature the iron undergoes allotropic transformation into beta iron. There is no further change until 1400

<sup>90</sup>The statement assumes that the cementite does not break down into iron graphite.

<sup>91</sup>See TRANSACTIONS, American Society for Steel Treating, February, 1927, Table I, page 292.



degrees Fahr. is reached at which point the beta iron is transformed into alpha iron. From 1400 degrees Fahr. the metal cools unchanged to atmospheric temperature when the metal is all in the form of pure alpha ferrite.

2—At 3000 degrees Fahr. a 0.20 per cent carbon steel is molten (ordinate L, Fig. 44), and consists of a molten solution of 3 per cent iron carbide in iron. Cooling is uniform until 2750 degrees Fahr. is reached (point  $L^1$ , Fig. 44) where the melt starts to solidify by depositing a solid solution of iron carbide in gamma iron known as austenite. This deposition of austenite continues while the metal is cooling from 2750 degrees Fahr.,  $L^1$ , to 2550 degrees Fahr.,  $L^2$ , at which temperature solidification is complete. Cooling is uniform to a temperature of 1550 degrees Fahr., point  $L^3$ , where part of the gamma iron is changed to beta iron enriching the remaining austenite in carbon. At 1400 degrees Fahr., point  $L^4$ , the beta iron is transformed into alpha iron. This transformation continues until 1300 degrees Fahr., point  $L^5$ , is reached when the remaining austenite, now of eutectoid composition, is changed to pearlite. At atmospheric temperature the steel is made up of 76.6 per cent free ferrite and 23.4 per cent pearlite.

3—At 3000 degrees Fahr. a 0.40 per cent carbon steel is molten (ordinate J, Fig. 44) and consists of a molten solution of 6 per cent iron carbide in iron. Cooling is uniform until a temperature of 2670 degrees Fahr. is reached (point  $J^1$ , Fig. 44). Here solidification starts by deposition of solid austenite. Solidification continues through the mushy stage  $J^1$  to  $J^2$  (2670 degrees Fahr. to 2460 degrees Fahr.). The now solidified austenite cools unchanged until a temperature of 1430 degrees Fahr. is reached. This temperature marks the beginning of the allotropic change of the gamma to beta and alpha iron which is complete at 1300 degrees Fahr., point  $L^4$ , Fig. 44. By the time this change is complete the remaining austenite has reached the eutectoid composition and transforms into pearlite. At atmospheric temperature the steel contains 46.8 per cent pearlite and 53.2 per cent free ferrite.

4—At 3000 degrees Fahr. a 0.90 per cent carbon steel is molten (ordinate H, Fig. 44) and consists of a molten solution of 13.5 per cent iron carbide in iron. Cooling is uniform until



a temperature of 2660 degrees Fahr. is reached (point  $H^1$ , Fig. 44). This is the liquidus where freezing of the austenite begins. Solidification is complete when the alloy has cooled to 2230 degrees Fahr. (point  $H^2$ , Fig. 44). The austenite cools unchanged from  $H^2$  to point S, Fig. 44, that is from 2230 degrees Fahr. to 1300 degrees Fahr. At this latter temperature the solid solution which is already of eutectoid composition changes to pearlite. At atmospheric temperature the steel is 100 per cent pearlite.

5—At 3000 degrees Fahr. a 1.45 per cent carbon steel is molten (ordinate K, Fig. 44) and consists of a molten solution of 21.7 per cent cementite in iron. Cooling is uniform until a temperature of 2600 degrees Fahr. is reached (point  $K^1$ , Fig. 44). Solidification starts at this temperature and is complete at 2100 degrees Fahr. (point  $K^2$ , Fig. 44). The solid solution cools unchanged until 1880 degrees Fahr. is reached (point  $K^3$ ) when precipitation of the excess cementite begins. This precipitation continues while the metal is cooling from 1880 to 1300 degrees Fahr. At this latter temperature (point  $K^4$ ) the remaining austenite, now of eutectoid composition changes to pearlite. At atmospheric temperature the steel is made up of 10 per cent cementite and 90 per cent pearlite.

6—At 3000 degrees Fahr. an iron-carbon alloy containing 3 per cent carbon (ordinate M, Fig. 44) is wholly molten and consists of a molten solution of 45 per cent iron carbide in iron. The alloy cools unchanged until a temperature of 2325 degrees Fahr. is reached (point  $M^1$ , Fig. 44). At this point solidification begins by the deposition of saturated austenite (a solid solution of 25.5 per cent cementite in gamma iron). As the temperature falls from 2325 degrees Fahr., point  $M^1$ , to 2065, point  $M^2$ , more saturated austenite is precipitated from the melt, enriching the remaining molten alloy in cementite. By the time the temperature has fallen to 2065 degrees Fahr. the remaining molten metal has reached the eutectic composition.

When solidification is complete the alloy is composed of 50 per cent excess or proeutectic austenite and 50 per cent eutectic, which is made up of 24 per cent eutectic austenite and 26 per cent eutectic cementite. As the alloy cools from 2065 degrees Fahr., point  $M^2$ , to 1300 degrees Fahr.,  $M^3$ , (the transformation range) the austenite rejects the cementite which is in excess of

the eutectoid ratio. When 1300 degrees Fahr. is reached this rejection is complete and the remaining austenite now of eutectoid composition is transformed into pearlite. At atmospheric temperature the alloy is made up of 64 per cent pearlite, 10 per cent proeutectoid cementite, and 26 per cent eutectic cementite. There are three classes of cementite present; (a) pearlite cementite, (b) proeutectoid cementite, (c) eutectic cementite.

7—At 3000 degrees Fahr. an iron-carbon alloy containing 4.30 per cent carbon (ordinate N, Fig. 44) is wholly molten and consists of a molten solution of 64.5 per cent iron carbide in iron. The alloy cools unchanged until a temperature of 2065 degrees Fahr. is reached (point B, Fig. 44) at which constant temperature the whole mass solidifies. After solidification the alloy is composed wholly of eutectic which is made up of 48 per cent saturated austenite and 52 per cent cementite. During the cooling from 2065 degrees Fahr. (point B) to 1300 degrees Fahr. (point N<sup>1</sup>) the saturated austenite rejects cementite until when 1300 degrees Fahr. is reached the austenite is of eutectoid composition and at this temperature is transformed into pearlite. At atmospheric temperature the alloy is made up of 41 per cent pearlite, 6.6 per cent proeutectoid cementite and 52 per cent eutectic cementite.

8—At 3000 degrees Fahr. an iron-carbon alloy containing 4.80 per cent carbon (ordinate P, Fig. 44) is wholly molten and consists of a solution of 72.0 per cent cementite in molten iron. The alloy cools unchanged until a temperature of 2550 degrees Fahr. is reached. At this temperature the excess cementite starts to solidify. This solidification continues while the temperature falls from 2550 degrees Fahr. to 2065 degrees Fahr. (point P<sup>1</sup> to P<sup>2</sup>). At this latter temperature the molten metal is of eutectic composition and freezes. As the alloy cools from 2065 degrees Fahr. to 1300 degrees Fahr. the austenite rejects the proeutectoid cementite until at 1300 degrees Fahr. the remaining austenite now of eutectoid composition changes into pearlite. At atmospheric temperature the alloy is made up of 32 per cent pearlite, 5.2 per cent proeutectoid cementite, 41 per cent eutectic cementite and 21 per cent excess cementite. Four classes of cementite are present: (a) the cementite in the pearlite aggregate, (b) the proeutectoid cementite which was rejected by the austenite in cooling, (c) the

eutectic cementite which together with the austenite made up the eutectic, and (d) the proeutectic cementite or the cementite in excess of the eutectic ratio.

The whole discussion of the iron-carbon diagram up to this point has considered the steels and cast irons as alloys of iron and iron carbide, in other words, ferrite and cementite. Cementite is not the most stable form in which carbon exists; as has been noted in a previous installment it tends to dissociate into iron and carbon. When this occurs we have (in the case of the higher carbon alloys) free carbon or graphite present. The iron-graphite alloys will be the next ones considered.

After we have discussed the constitution of the gray irons, we will proceed to the discussion of the influence of the other elements, manganese, silicon, sulphur and phosphorus on the cast irons. This completed we will return to the steels and view their behavior when they are in a state of unstable (or metastable) equilibrium, in other words, when cooling is accelerated when the alloys are passing through the transformation ranges. This will constitute an elementary discussion of the theory of heat treatment.

## Comment and Discussion

Papers and Articles Presented Before the Society and Published in Transactions Are Open to Comment and Discussion in This Column

### DISCUSSION OF DR. V. N. KRIVOBOK'S PAPER ENTITLED "DENDRITIC CRYSTALLIZATION AND GRAIN FORMATION IN STEELS"

BY DR. ALBERT SAUVEUR

DR. KRIVOBOK, in his interesting contribution to the crystallization of iron and of steel, offers some valuable evidence in support of the contention that the dendrites formed during solidification break up into small micrograins on subsequent cooling. It does not seem reasonable to assume that the allotropic transformation iron undergoes when it passes from its gamma to its alpha condition is responsible for this granulation, in view of the evidence we have that other metals and alloys which do not undergo allotropic transformation exhibit dendritic granulation. Noting Dr. Krivobok's statement that, for like subsequent treatment, steel originally coarsely dendritic does not necessarily exhibit a coarser microstructure than a steel originally finely dendritic; and, admitting the correctness of that view, it does not, of course, follow that the size of the dendrites produced on solidification does not affect the final properties of manufactured articles. It remains true that large dendrites promote banding and directional properties, and that this is more marked in impure steels and in steels containing an excessive amount of inclusions. This is in line with the results obtained by Mr. W. P. Benter, reported by Dr. Krivobok, in casting steel in molds respectively hot and cold. The marked banding of the steel cast in the hot mold resulted undoubtedly from the large dendrites produced, which also implies more intense dendritic segregation. Even in chemically pure iron-carbon alloys, banding will occur because of the dendritic segregation of carbon, although it should be less pronounced than in impure alloys. Slow solidification, implying as it does the formation of large dendrites, should promote this banding. Dr. Krivobok's photomicrograph No. 25 is of especial interest, as it shows that granulation has taken place *after* the steel had been "banded" by hot work.

Dr. Krivobok is to be congratulated on the skill with which his experiments were conducted and the soundness of his conclusions.

<sup>1</sup>TRANSACTIONS of American Society for Steel Treating, Vol. 10, November, 1926, pages 758 to 781. Also Preprint No. 6, eighth annual convention, Chicago, September 20 to 24, 1926.

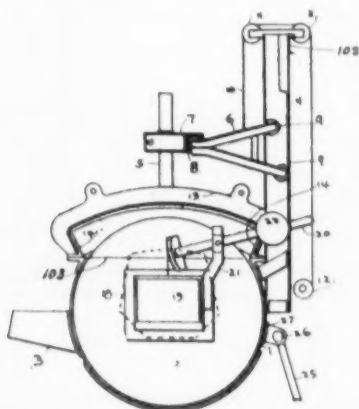


## Reviews of Recent Patents

By NELSON LITTELL, Patent Attorney  
475 Fifth Ave., New York City—Member of A. S. S. T.

### 1,616,796, Electric Furnace, A. E. Greene, of Seattle, Washington.

This patent describes an electric furnace comprising a cylindrical body portion 1, a discharge spout 3, a separate top member 13, end doors 19, and electrodes 5 supported from vertical eye beams 4 by means of the trolley



members 6 which are slidable vertically from the eye beams, 4 by means of the rollers 9 and are controlled by the cable 10. The electrode holder 7 is pivotally connected to the outer end of the trolley 6 so as to permit adjustment of the electrode either horizontally or vertically.

### 1,619,025, Heat Treatment of Steel, Edward F. Kenney, of Bethlehem, Pa.

This patent relates to a process of heat treating steel, particularly, railway rails, wheels, and the like which have different degrees of thickness. As an example of the invention, rails of the composition—

	Per Cent
Carbon	0.63 to 0.87
Manganese	0.60 to 0.90
Phosphorus	0.03
Sulphur	0.05
Silicon	0.15 to 0.25

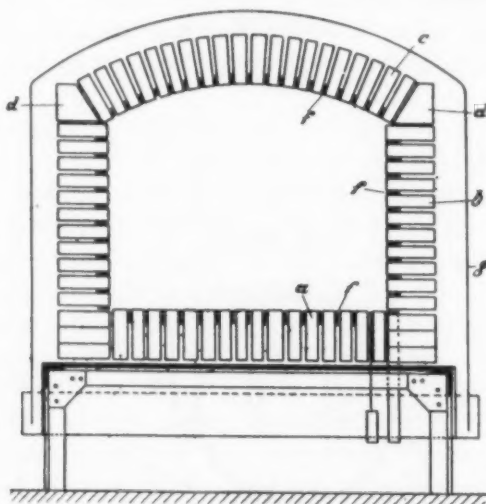
were quenched in water for about thirty seconds from a temperature about 1750 degrees Fahr., bringing the temperature down to about 725 degrees Fahr., i. e. from above a temperature above the critical range to one below the critical range, above the blue heat range. The rails were then charged into a furnace maintained at 1000 degrees Fahr., and were kept at this temperature for one hour, removed and allowed to cool in air. The re-heat-



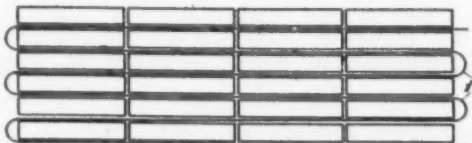
ing after quenching was for the purpose of equalizing the heat in the various sections and the rails produced in this manner were much tougher under tests than rails treated in the usual manner.

1,619,829, **Electric Resistance Muffle Furnace**, Wilhelm Rohn, of Hanau-on-the-Main, Germany, assignor to Charles Engelhard, Inc., a corporation of New York, N. Y.

This patent describes an electric resistance furnace in which the resistance members *F* are formed in substantial continuous bands of flat



material imbedded in the spaces between the bricks *a* of the bottom, *b* of the side walls and *c* of the top being zigzagged back and forth between the



crevices of the bricks. The electrical connections in this manner may be reduced to a minimum and a furnace constructed in an economical manner.

1,619,399, **Production of Thin Steel**, Porter H. Brace, of Wilkesburg, Pennsylvania, assignor to Westinghouse Electric & Manufacturing Company, a corporation of Pennsylvania.

This patent describes the process of producing extremely thin steel sheets for use in cores and pole pieces of high frequency electrical apparatus. It is the object of the invention to provide sheets of a thickness of 2 mils or less having a smooth surface and uniform cross-section. The extreme thinness is produced by rolling a pack of thicker sheets at a temperature of 1380 to 1470 degrees Fahr. (750 to 800 degrees Cent.) to produce individual sheets of 5 mils in thickness which are then individually cold-rolled, then re-stacked and hot-rolled at a temperature of 1290 to 1380 degrees Fahr. (700 to 750 degrees Cent.) to produce sheets of about 3 mils

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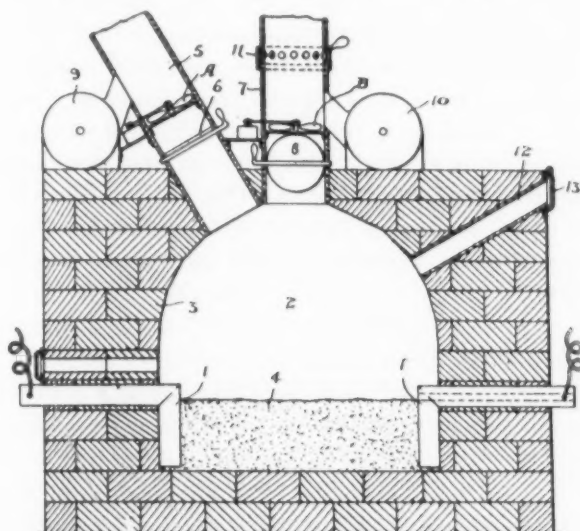
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in thickness which are again individually cold-rolled, re-stacked and annealed at 1290 to 1380 degrees Fahr. (700 to 800 degrees Cent.), and cold-rolled in the pack to produce sheets of about 1.5 mils in thickness. Further reduction in thickness may be made by carrying the process further.

**1,620,880, Process of Producing Metallic Oxides, George B. Filmer, of Toronto, Ontario, Canada, assignor of One-Half to Walter James Harvey, of Toronto, Ontario, Canada.**

This patent describes a furnace for producing metallic oxide composed of a refractory bottom 4 through which the ends of the electrodes 1 extend. 3 indicates the walls of the furnace; 2, the furnace chamber, 12 the chute through which the metal to be oxidized is introduced, and 5 and 7 flues for



controlling the exhaust and inlet of gases into the furnace. In the use of the furnace, zinc for example is fed from the chute 12 on to the bath of the molten refractory material 4 and as the fumes rise through the flues 7 air is admitted in regulated means by means of the air passages 11 to convert the fumes in an oxide which is precipitated into any suitable receptacle. The flue 5 is used to remove the waste gases from the furnace when operation is first started and before the metal is introduced therein.

**1,620,081 and 1,620,082, Alloy of Lithium and Aluminum, Johann Czochralski and George Welter of Frankfort-on-the-Main, Germany, assignors, by Mesne Assignments, to Allied Process Corporation, of New York, N. Y., a corporation of New York.**

These patents describe alloys of lithium and aluminum containing also zinc, copper, manganese, and other materials if desired. The aluminum lithium alloy may contain as much as 40 per cent of lithium and should be substantially free from silicon. The alloy may be used for drawing, rolling, forging, welding, and the like, and may be substantially improved by heat treatment.

## Abstracts of Technical Articles

### Brief Reviews of Foreign Publications of Interest To Metallurgists and Steel Treaters

FUSION WELDING AND ENDURANCE FRACTURE. By Dr. H. Schottky. *Kruppsche Monatshefte*, December, 1926, p. 213 to 216.

This paper deals with fractures of machine parts, being connected with the applying of welding material and reasons are discussed why this repairing method exposes the parts to the liability of failure.

In all cases the fracture proceeds from the weld and as the applied material possesses only a low carbon content and numerous inclusions of slag and pores, cracks are formed in it when the pieces are put under endurance stress; these cracks act as notches in the material and the danger of breaking is considerably increased; furthermore the original material undergoes a change on its surface by the welding, which may be called as overheating or feeble hardening. Greater attention should be paid to the application of welding.

*Abstracted by Dr. Hans Pollack, Germany.*

THE PRESENT STATE OF HARD METALS AND THEIR APPLICATION. By Ing. E. Güldner, Tübingen. *Maschinenbau*, August 19, 1926, p. 744-746.

The causes of the failures in the use of American stellite are founded on its unequalness, on its porosity and on the difficulty experienced in joining the alloy with the support by soldering or welding.

Today these deficiencies have been overcome. Different German firms produce a hard metal of uniform chemical composition, practically free from pores.

By the application of Walter's patent there is a marked improvement in the junction of the hard alloy with its support through the elimination of pores. This invention provides that the hard metal be deposited on the support in the liquid state.

The formation of pores may be entirely avoided by skilful handling of the method. The cutting qualities of the hard metal are improved by the grain refining taking place.

Hard metal tools may be employed for rough machining as well as for smoothing. For small cutting effects the hard metal is unsuitable and inferior to high speed steel. The cutting speed depends on the properties of the material being cut and on the cross section of the turnings.

The speed of machining is dependent upon the tensile strength of the material, although the toughness is of great importance.

Brass, copper, bronze may be economically machined with hard metal; the high cutting speed of 675 to 975 feet per minute, required for this machin-

ing, may only be obtained in few cases on the commonly used machines.

The application range of hard metal extends from 300 to 600 degrees Cent. (572 to 1112 degrees Fahr.) and even to 800 degrees Cent. (1472 degrees Fahr.).

*Abstracted by Dr. Hans Pollack, Germany.*

THE INCREASE OF THE DURATION OF LIFE OF DIES FOR HOT-STAMPING BRASS. By Dr.-Ing. A. Aronheim, Berlin. *Maschinenbau*, October 7, 1926, p. 887 to 892.

This paper deals with the investigations made by the author in the Oberspree cable work of the AEG to clear up the causes of the great die consumption and devise a method of increasing the durability of dies.

The author discusses detailed the qualification of the machines used for the hot stamping of brass and the properties of die steel.

Carbon steel, containing 0.9-1.00 per cent carbon and case-hardened open-hearth steel possess a low durability for this purpose.

Then the author used a chromium-nickel steel of the following chemical composition: carbon, 0.16; nickel, 4.72; chromium, 1.41; manganese, 0.46; silicon, 0.31; phosphorus, 0.03; sulphur, 0.02. This steel was heated on 800 degrees Cent. (1472 degrees Fahr.) and air-cooled or quenched in oil and tempered at 250 to 300 degrees Cent. (482 to 572 degrees Fahr.), but the so treated dies inclined to cracks.

This chromium-nickel steel was case-hardened; the carburizing temperature was 750 degrees Cent. (1382 degrees Fahr.). After carburizing the die was cooled in the box out of the furnace, reheated to 750 degrees Cent. (1382 degrees Fahr.), quenched in oil and drawn at 300 degrees Cent. (572 degrees Fahr.). When so treated and heated the best results were obtained. The durability of the dies was increased by the three-to-fivefold according to the form of the die.

*Abstracted by Dr. Hans Pollack, Germany.*

APPLICATION OF X-RAYS FOR THE EXAMINATION OF MATERIALS. By G. Sachs, Berlin-Dahlem. Report of the Kaiser-Wilhelm-Institut fuer Metallforschung. *Zeitschrift des Vereins Deutscher Ingenieure*, December 4, 1926, p. 1634 to 40.

The author gives a summary of the application of X-rays in the science of metals and the results of practical value gained by it.

The author starts his discussion with the reflection phenomena of ordinary light on crystal facets. A crystal consists of a space lattice of atoms, succeeding one another periodically in the three directions of the space. In most cases the crystals are imperfectly formed, and in such cases it is often possible to lay bare natural facets by clearing. Single crystals of metals are manufactured in different forms, such as rods, wires or sheets. By suitable etching, reflection phenomena may be revealed on these crystals, making a determination of the orientation possible. Such determinations have proven that crystals select in their growth important lattice directions as lines of growth.



Then the author describes the Laue-method. A photogram after Laue is made by lighting up an unbroken crystal with a sheaf of X-rays of different length of undulation; the diffracted rays fall on a photographic plate set up vertical to the original sheaf of X-rays. When the incidence X-rays parallel an important line of the crystal, then the photogram is regular, otherwise irregular. At known lattice it is possible to determine its position in the sample from the Laue-photogram.

By cold deformation the metal crystal is divided by a great number of parallel gliding planes in which its single parts get out of their places. In most cases the deformation is connected with a turning of the crystal lattice, opposite to the directions of the stress and crookedness of the single crystal parts. The changes in the X-ray photograms are also interpreted by the deformation. The examination of single crystals promises to throw light on the recrystallization and hot-deformation proceedings, being of great practical value. In the investigation of metals, monochromatic Röntgen-light is the one most used. From the position and intensity of the interference circles on a film it is possible to determine not only the number of the present sorts of crystals, but also their composition in all detail. The author discusses briefly the determination of the lattice-type and its constants from a Debye-Scherrer photogram. From the nature of the crystal types it is possible to predict the properties of the alloys. Another question of great importance is the transformation points of the alloys, e. g., the modifications of iron and the hardening procedure. In aluminum-alloys X-ray examination has proved, hitherto, only the absence of greater structural changes. Cold deformed metals change the X-ray photograms. By annealing, the metals recrystallize and the beginning and the progress of the recrystallization may be indicated just as certainly as with the aid of the metallographic examination. For examinations where the test piece cannot be destroyed, such as boiler plates, the X-ray examination is of great use. In the foundry and in the electrolysis of metals the X-ray examination may obtain possibly further practical value. The author further discusses the application of X-rays in the chemical analysis and refers to the emission and to the absorption method. In qualitative analysis, X-ray examination has worked out well as the discovery of Hafnium and of other elements prove; elements with a great atom-weight may be indicated in the smallest quantity, but the determination of elements with a small atom-weight is difficult; quantitative methods will be developed. The lighting up with X-rays will become a useful testing method in foundry and welding practice. Materials possessing a small atom-weight, as aluminum, may be easily lighted up, while those of great atom-weight may be lighted with difficulty. The present limits of the thickness of the sample up to which e. g. inclusions of air of 1 millimeter may still be proved with certainty lies below 10 centimeters for steel and 8 centimeters for brass. The author finally describes briefly the layout of an X-ray laboratory and comes to the conclusion that the method must be developed and the apparatus must be simplified so that the smaller industry laboratories may make use of it with profit.

*Abstracted by Dr. Hans Pollack, Germany.*

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# THE ENGINEERING INDEX

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Arrangements have been made with The American Society of Mechanical Engineers whereby the American Society for Steel Treating will be furnished each month with a specially prepared section of The Engineering Index. It is to include items descriptive of articles appearing in the current issues of the world's engineering and scientific press of particular interest to members of the American Society for Steel Treating. These items will be selected from the copy prepared for the annual volume of the Index published by the A. S. M. E.

In the preparation of the Index by the staff of the A. S. M. E. some 1,200 domestic and foreign technical publications received by the Engineering Societies Library (New York) are regularly searched for articles giving the results of the world's most recent engineering and scientific research, thought, and experience. From this wealth of material the A. S. S. T. will be supplied with a selective index to those articles which deal particularly with steel treating and related subjects.

Photostatic copies (white printing on a black background) of any of the articles listed may be secured through the A. S. S. T. The price of each print, up to 11 by 14 inches in size, is 25 cents. Remittances should accompany orders. A separate print is required for each page of the larger periodicals, but whenever possible two pages will be photographed together on the same print. When ordering prints, identify the article by quoting from the Index item: (1) Title of article; (2) name of periodical in which it appeared; (3) volume, number, and date of publication of periodical; and (4) page numbers.

## ALLOY STEELS

**ELECTROTECHNICAL.** Special Steels (Les aciers speciaux), M. R. Jouaust. Societe Francaise des Electriciens—Bul., vol. 6, no. 62, Oct., 1926, pp. 1175-1182. Discusses magnetic substances of high initial permeability and small total losses, particularly with reference to their use in long-distance telephone lines; magnetic data are given for iron-dust cores made by Western Electric Co., and mode of preparation of these cores, which may advantageously be used for Pupin coils; composition of permalloy; alloy, with its high permeability in weak fields, small hysteresis losses and high resistivity, is particularly suitable for use in Krarup process, thin ribbon of this material being wound round cable; permalloy, a nickel-steel alloy is mentioned; account of cobalt steels of high coercive force, such as KS alloy of Honda, for making permanent magnets.

**HIGH-TENACITY.** Cheap High-Tenacity Steels. Machy. (Lond.), vol. 29, no. 745, Jan. 20, 1927, p. 529. Presents table giving mechanical properties of various steels after heat treatment; shows that high-manganese carbon steel is cheaper than 3 per cent nickel steel, and is not much more expensive than plain carbon steel.

## ALLOYS

**HEAT-RESISTING.** Use of Heat Resisting Alloys in Construction of Ore Roasting Furnaces. Fuels and Furnaces, vol. 5, no. 2, Feb., 1927, pp. 235-236, 2 figs. Hollow rubble arms made of heat-resisting alloy function as recuperators, giving increased furnace efficiency.

## ALUMINUM

**ELECTRICAL INDUSTRY.** Aluminum in the Electrical Industry (Aluminium in der Elektrotechnik), W. Wunder. Zeit. für Metallkunde, vol. 19, no. 1, Jan., 1927, pp. 36-37, 1 fig. Importance of aluminum in

electrical engineering is based on comparatively high electrical conductivity, light weight of metal, and good mechanical properties of its alloys; use of pure aluminum in motors, generators and transformers; uses of aluminum alloys; in author's opinion, however, little progress has been made in this field in past few years.

**GERMAN INDUSTRY.** The Aluminum-Product Industry in Germany (Die Aluminium-Fertigindustrie Deutschlands), R. Görnandt. Zeit. für Metallkunde, vol. 19, no. 1, Jan., 1927, pp. 38-40. Aluminum and its field of usefulness; present status and future prospects; of industry for aluminum household ware, industrial products, apparatus, castings; export problems.

**GERMAN PLANTS.** Aluminum Works in Germany (Die Aluminiumwerke in Deutschland), Menzen. Zeit. für Metallkunde, vol. 19, no. 1, Jan. 1927, pp. 4-5, 4 figs. Establishment and present status of German industry, with brief descriptions of different plants.

**ORGANIC-CHEMISTRY INDUSTRY.** The Use of Aluminum in Organic Chemistry (Aluminium als Werkstoff in der organischen Chemie, Buschlinger. Zeit. für Metallkunde, vol. 19, no. 1, Jan., 1927, pp. 25-36, 6 figs. Review of experimental results in chemical industry, giving, it is claimed, for first time systematically arranged basic requirements for use of light metals; tests applied to coal-gas, petroleum, fermentation, fatty-acid, sugar, foodstuff and other industries.

**PRODUCTION METHODS.** The Technical Production of Aluminum (Die Technische Darstellung des Aluminiums), W. Fulda. Zeit. für Metallkunde, vol. 19, no. 1, Jan., 1927, p. 3. Brief review of methods heretofore employed and reference to newer methods and possibilities.

**PROGRESS.** A Century of Aluminum, H. Groeck. Eng. Progress, vol. 8, no. 2, Feb., 1927, pp. 39-40, 3 figs. Review of developments undergone by aluminum and partic-

ulars of its high-grade alloys. (Abstract.) Translated from *Zeit. für Metallkunde*.

**SOLDERING AND WELDING.** Some Practical Notes on Soldering and Welding Aluminum, A. Eyles. *Engineer*, vol. 143, no. 3708, Feb. 4, 1927, pp. 121-123, 5 figs. It is shown that aluminum can be welded but it is not advisable to employ soldering methods, strictly speaking, for assembly of structures; its only actual feasible applications are for repairing fractured or broken parts, and for filling up surface holes and slight defects in castings; it can also be applied on stressed members or on any components, failure of which might be serious; two distinct methods used in welding aluminum are puddling and flux systems; preparation of metal for welding and execution of welds in aluminum.

**WORLD PRODUCTION.** The World Production of Aluminum since its Discovery (Die Welterzeugung an Aluminium seit seiner Entdeckung), E. Scheuer. *Zeit. für Metallkunde*, vol. 19, no. 1, Jan., 1927, pp. 6-8, 1 fig. Relation between technical progress in methods of production and extent of production as well as costs; experimental results of St. Claire-Deville; tests in England and Germany; influence of discovery of electrolytic process on cost and extent of production; present status of industry in world and in Germany.

#### ALUMINUM ALLOYS

**AIRPLANE MATERIAL.** Light Metal for Airplane Construction (*Lättmetall für flyplanskonstruktioner*), T. Angström. *Ingeniörs Vetenskaps Akademien—Handlingar*, no. 51, 1926, pp. 5-30, 13 figs. Deals with ultra-light alloys, light aluminum alloys capable of being rolled, including aludur, lautal, skleron and duralumin; heat treating and working of duralumin, corrosion. Bibliography.

**ATMOSPHERIC EXPOSURE, EFFECT OF.** The Corrosion Products and Mechanical Properties of Certain Light Aluminum Alloys as Affected by Atmospheric Exposure, E. Wilson. *Physical Soc.—Proc.*, vol. 39, no. 216, Dec. 15, 1926, pp. 15-25, 3 figs. Experiments made on electrical conductivities, corrosion products and tensile properties of high-purity aluminum and certain light aluminum alloys, which have been exposed to London atmosphere for period of 24 years; elements concerned are copper, nickel, manganese and zinc in varying amounts up to few per cent; includes note on corrosion products of high-conductivity copper.

**DEFORMATION.** Deformation of an Aluminum Alloy by a Constant Load, C. B. Sadtler and J. L. Gregg. *Am. Inst. Min. & Met. Engrs.—Trans.*, no. 1645-E, Mar., 1927, 6 pp., 4 figs. Deals with variations of deformation produced by constant tensile stress acting for varied lengths of time in special case of thin aluminum alloy sheet of duralumin type, which has been subjected to different thermal and mechanical treatments; it was found that at sufficiently high constant tension such material deforms continuously with elapse of time and that rate at which this deformation occurs is greatly influenced by previous thermal and mechanical treatment.

**HIGH-GRADE.** Aluminum Alloys Capable of Refining (Die veredelbaren Legierungen

des Aluminiums). *Zeit. für Metallkunde*, vol. 19, no. 1, Jan., 1927, pp. 9-21, 20 figs. Contains following contributions: Recent Developments in Refining Processes, Especially with Aluminum-Alloy Castings, K. L. Meissner; Duralumin, R. Beck; Silumin, and Refining Processes, J. Czocharski; Scleron and Aeron, E. Scheuer; Lautal, Fuss; Constructal, W. Sander.

**LAUTAL.** The Wrought Light Alloy "Lautal". *Engineering*, vol. 123, no. 3186, Feb. 4, 1927, p. 133. This alloy, which has recently come into prominence, possesses property of age hardening, but does not age automatically at room temperatures; special tempering treatment is necessary in order to produce maximum tensile strength and Brinell hardness obtainable; this is main advantage claimed by originators of alloys; aging is carried out on finished articles, after forging and working operations are completed, process consisting of heating them in oil bath for 16 hours at temperature of 120 to 130 deg. Cent.; it contains 4 per cent copper, 2 per cent silicon and remainder is made up of aluminum of commercial purity.

**PROPERTIES AND USES.** Aluminum Alloys, O. Hyman. *Foundry Trade Jr.*, vol. 35, no. 545, Jan. 27, 1927, pp. 74-77, 1 fig. Nature of aluminum; commercial impurities; effect of alloy in copper; aluminum-silicon alloys; commercial alloys; improving properties; corrosion; heat treatment; aging; foundry application; gating; contraction. In discussion, action of chills, American practice compared, and aluminum alloys at elevated temperatures are touched upon.

**TRANSPORTATION EQUIPMENT.** Light Metal in Transportation Equipment (Das Leichtmetall im Verkehrswesen), H. Kühnen. *Zeit. für Metallkunde*, vol. 19, no. 1, Jan., 1927, pp. 22-24, 7 figs. Discusses increasing use of high-grade aluminum alloys in automobile construction, cycles and motor-cycles, rolling stock, aircraft and shipbuilding; advantages include increase of starting acceleration, better braking efficiency, increased average speed, greater safety, less power requirement, and greater economy.

#### AXLES

**STREET-CAR.** Tramcar Axles. *Tramway & Ry. World*, vol. 61, no. 8, Feb. 17, 1927, pp. 81-83, 1 fig. Problems of axle design; investigation of joint committees on rolling stock; recent developments; effects of correct heat treatment; correlation of various tests; required properties of axle steel.

#### BEARING METALS

**BABBITT.** Proper Selection of Babbitt Metal Important, C. W. Hudson. *Belting*, vol. 30, no. 2, Feb., 1927, pp. 36-42. Chief considerations in design, casting and fitting of babbitt bearings; service requirements should always be considered in selection of metals; preparations for casting; gating and venting; casting bearing.

**CALCULATING METAL QUANTITIES.** Calculating Metal Quantities for Alloys, A. W. Beshgetoor. *Am. Mach.*, vol. 66, no. 8, Feb. 24, 1927, p. 331, 1 fig. Method suitable for calculating bearing-metal alloys to any desired composition.

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## BLAST FURNACES

**COMPOSITION OF MATERIALS FROM.** Composition of Materials from Various Elevations in an Iron Blast Furnace, S. P. Kinney. U. S. Bur. of Mines—Tech. Paper, no. 397, 1926, 22 pp., 6 figs. Investigation of gas composition, temperature and pressure at series of planes in blast furnace making foundry iron in southern district; samples of coke, metal, slag, limestone, and stock were recovered from points between tuyere plane and stock line; aside from giving position of zones of reduction, calcination and slag formation, samples would be valuable in determining where iron acquires its carbon, sulphur, phosphorus, manganese and silicon.

**EXPERIMENTAL OPERATION.** Experimental Blast Furnace Operation. Min. & Met., vol. 8, no. 243, Mar., 1927, pp. 126-127, 5 figs. Review of blast furnace investigation undertaken by Bureau of Mines in 1919.

**GREAT BRITAIN.** Blast Furnaces of the United Kingdom Built and in Blast. Foundry Trade J. (Supp.), Jan. 27, 1927. Statistics for quarter ending Dec. 31, 1926.

**PRODUCER GAS.** Partial or Total Replacement of Metallurgical Coke by Producer Gas in Blast Furnace (Du remplacement partiel ou total dans les hauts fourneaux du coke metallurgique par du gaz de gazogene), R. F. Lance. Chime & Industrie, vol. 16, no. 3, Sept., 1926, pp. 399-405. Shows that use of producer gas permits elimination in charge of at least 70 per cent of coke, which can be replaced by cheaper fuels; proposed system requires 2 rows of tuyeres about 1 m. apart, 2 sets of recuperators, and one or more producers burning coal or lignite.

**REACTIONS.** Contribution to Thermodynamics of Blast-Furnace Process (Ueber das Gleichgewicht  $C + CO_2 \rightleftharpoons 2CO$ , ein Beitrag zur Thermo dynamik des Hochofenprozesses), V. Falcke. Zeit. für Elektrochemie, vol. 33, no. 1, Jan., 1927, pp. 1-11, 20 figs. In a study of thermodynamics of blast-furnace ore reduction, good agreement was found between theory and practice; measurements at 750 to 1000 degrees Cent. agreed with calculated effects of heat of reaction (combustion of graphitic carbon and of CO) and specific heats of substances involved; equilibrium of CO with carbon and  $CO_2$  was studied in particular.

**REDUCTION BY GAS.** New Blast Furnace Process (Ein neuer Hochofen prozess), P. W. Uhlmann. Chemiker-Zeitung, vol. 51, no. 4, pp. 37-38. From theoretical considerations (thermal relations), new process of using gas for blast-furnace reduction has some advantages over coke reduction and should be commercially successful; most satisfactory gas is water gas enriched with hydrogen to contain about 35 per cent CO and 65 per cent  $H_2$ .

**SPECIFIC EFFICIENCY.** Specific Efficiency of the Blast Furnace, R. Franchot. Am. Inst. Min. & Met. Engrs.—Trans., no. 1596-C, Sept., 1926, 18 pp., 1 fig. Presents analysis showing that typical furnace is not less efficient as gas producer than as iron smelter; improvement of efficiency in iron making is to be sought through utilizing some of latent energy of  $CO_2$  which now passes undeveloped and unused out of furnace top; it narrows problem down to one of increasing  $CO_2$  ratio, thus developing greater proportion of energy in form of coke and air put into furnace; efficiency of Lowthian Bell's furnace; relation of energy input to work done; surplus energy in coke furnaces.

## BOILER PLATE

**EMBRITTEMENT.** The Embrittlement of Boiler Plate, F. G. Straub. Blast Furnace & Steel Plant, vol. 15, no. 2, Feb., 1927, pp. 94-99 and 107, 10 figs. Reasons for formation of cracks; micrographic studies aid in determining character of cracks; causes and prevention of embrittlement.

## BOILER TUBES

**DAMAGES.** Damages to the Tubes of Small-Tube Water-Tube Boilers, J. Levai. Am. Soc. Naval Engrs.—Jl., vol. 39, no. 1, Feb., 1927, pp. 39-51, 11 figs. Results of inspection of small-tube, water tube boilers and water pipes subjected to high stresses; tubes were made according to Mannesmann-Pilger process, in which, according to specifications, they were drawn through passes while cold; type of erosions displayed is indication of fatigue of metal.

**ROLLING.** Rolling Tubes in Boiler Plates, P. H. Oppenheimer. Power, vol. 65, no. 8, Feb. 22, 1927, pp. 300-303, 9 figs. Strength of rolled joints, power required to roll them, and effect of rolling on metal of both tube and sheet; process of rolling is composed of three operations: (1) Expanding tube until it touches wall of hole in which it is being fastened; (2) Rolling tube and deforming metal of plate surrounding it; (3) Withdrawing expander.

## BRASS

**DEFECTS.** Common Defects in Brass, Metallurgist (Supp. to Engineer), Feb. 25, 1927, pp. 29-30. Common defects encountered in brass articles made from sheet and strip may be divided into two distinct categories, those which are due to defects in metal as cast, and those which arise in course of subsequent operations; defects, such as red stains, occur during course of manufacture; defects of congenital origin show themselves as blisters, spills, laminations, cavities and dirty porous patches, all of which are associated with casting conditions; type of unsoundness which gives rise ultimately to blisters is largely obviated by pouring at high temperature; unsoundness due to shrinkage cavities is much more difficult to eliminate.

**HEAT TREATMENT AND HARDNESS.** A Study of the Heat Treatment, Microstructure and Hardness of 60:40 Brass, F. H. Clark. Am. Inst. Min. & Met. Engrs.—Trans., no. 1630-E, Mar., 1927, 30 pp., 91 figs. Experimental work to study effect of mechanical work on rolled 60:40 brass, quenched and reheated, and mechanism of twinning in alpha reeds of furnace-cooled specimen; it was desired to determine effect of mechanical work on transformations taking place on reheating quenched material, both at 200 and 470 degrees Cent.

**HOT ROLLED.** Hot Rolled Brass Bars. Soc. Mech. Engrs. (Tokyo)—Jl., vol. 30, no. 117, Jan., 1927, pp. 1-29, 55 figs. Great variation of quality of hot rolled brass bars



of same composition and of same manufacture can be detected by examining value of ultimate shearing strain or roughness of surface of test piece which it retains after testing; based on tension and torsion test and examination of inner structure of such materials, author has ascertained that this variation of quality is in accordance with temperature at which bars are presumed to have been rolled; tensile and torsional strength and measures of toughness to be obtained by tension test may fail to furnish any evidence of such variation of quality. (In Japanese).

#### BRASS FOUNDRIES

**FURNACE TYPES.** Furnace Types in Brass Foundries. *Brass World*, vol. 23, no. 2, Feb., 1927, pp. 49-51. Important features of considerable amount of data submitted by more than 80 brass foundries; furnace types and fuels; fluxes and linings receive careful analysis and consideration.

#### BRONZES

**GERMAN RAILWAY.** The New R 5 Bronze for German Railways (Der neue Rotgufs R 5 im Eisenbahnbetrieb), Kühnel and Marzahn. *Organ für die Fortschritte des Eisenbahnwesens*, vol. 82, no. 1, Jan. 15, 1927, pp. 11-14, 4 figs. Results of tests carried out by Railway Administration on copper-tin-zinc alloys to determine hardness, notch hardness, abrasion resistance, etc.; results show that this bronze possesses greater wear resistance than German standard R 9 bronze.

#### CASE-HARDENING

**NITRATION.** Nitration Hardening. Machy. (Lond.), vol. 29, no. 741, Dec. 23, 1926, pp. 393-395, 5 figs. New gas process for case-hardening steel parts developed by Krupps.

#### CAST IRON

**CUPOLA PRACTICE, INFLUENCE OF.** The Physical Structure of Cast Iron and its Relation to Cupola Practice, E. Howell. *Australasian Inst. Min. & Met.—Proc.*, no. 63, Sept., 1926, pp. 11-16, 7 figs. Points out that if cupola is not perfect melting medium, metal leaving spout has not sufficient superheating to give best results; casting temperature and rate of pouring, taken all in all, have bigger influences on production of close-grained castings than analysis; cupola dimensions.

**DIFFUSION IN.** The Diffusion of Sulphur into Cast Iron, F. Roll. *Foundry Trade Jl.*, vol. 35, no. 545, Jan. 27, 1927, pp. 83-84, 7 figs. Author describes experiments carried out to determine depth of penetration of sulphur into cast iron and its influence upon structure; he also studied action of solid sulphur  $H_2S$  and  $SO_2$ . Translated from *Giesserei*, vol. 14, no. 1, Jan. 1, 1927, pp. 1-7, 15 figs.

**ENAMELING.** A Note on the Enameling of Cast Iron. *Metal Industry* (Lond.), vol. 30, no. 8, Feb. 25, 1927, pp. 219-220. Defects in enameling; free graphite and blister formation.

**GRAPHITIZATION.** Influence of Carbon and Silicon on the Graphitization of White Castings (Influence du carbone et du silicium

sur la graphitisation des fontes blanches), P. Chevenard and A. Portevin. *Académie des Sciences—Comptes Rendus*, vol. 183, no. 25, Dec. 20, 1926, pp. 1283-1284, 1 fig. Presents 3-dimensional diagram to show influence of carbon and silicon on temperature of spontaneous graphitization of pure casting containing less than 0.1 per cent of manganese, sulphur or phosphorus; proportions of carbon and silicon are such as to include most industrial castings and hypoeutectics in iron-carbon system; curves for effect of silicon are approximately hyperbolic and indicate that beyond 2.5 to 3.5 per cent Si, graphitization temperature is almost constant at about 600 degrees.

**LOW-CARBON.** High Strength Cast Iron of Low Total Carbon Content. *Foundry Trade Jl.*, vol. 35, no. 545, Jan. 27, 1927, pp. 79-80. Synthetic and pseudo-synthetic irons; semi-steel cast iron; Thyssen-Emmel iron.

**MICROSCOPIC EXAMINATION.** Cast Iron Under Microscope. *Foundry Trade Jl.*, vol. 35, no. 547, Feb. 10, 1927, pp. 113-114 and 184. Discussion of paper by C. R. Groves; over-annealing and normalizing; silico-ferrite and carbon precipitation; inverse chill.

**NICKEL.** Nickel and Nickel-Chromium Cast Iron as Now Used in America, T. H. Turner. *Foundry Trade Jl.*, vol. 35, nos. 544 and 545, Jan. 20 and 27, 1927, pp. 59-61 and 71-73, 5 figs. Influence of Ni and of Ni-Cr additions to cast iron; machinable hardness; extent to which Ni-Cr alloy cast irons are now being used; automobile castings; Diesel-engine iron. Jan. 27: Marine cylinders; locomotive cylinders, liners and other cast iron parts; cost of making nickel and nickel-chromium additions to cast iron; application to centrifugal casting; method of adding nickel and chromium.

**OXYGENATED.** Oxygenated Cast Iron, T. Hill. *West. Machy. World*, vol. 18, no. 2, Feb., 1927, pp. 73-74. Results of tests on experimental plant in California operated by Triumph Steel Co. at Cristo, to determine feasibility of smelting briquetted magnetite concentrates, and also to determine feasibility of using briquetted concentrates in commercial foundry practice as means of oxygenating gray iron and to displace steel scrap to attain semi-steel physical properties.

**SILICON DETERMINATION.** Determination of Silicon in Pig Iron and Cast Iron (Beitrag zur Siliciumbestimmung im Roheisen und Guss), H. Pinsl. *Chemiker-Zeitung*, vol. 50, no. 112, Dec. 8, 1926, pp. 924-925. Results obtained for silicon in pig iron and cast iron by method of Rubricus are usually a little low unless filtrate from first precipitate is again evaporated; the longer the heating with strong sulphuric acid the more easily does precipitate filter; more accurate results with only one filtration are obtained if 20 cc. of 1:1 hydrochloric acid are added after first evaporation and evaporation is repeated, followed by heating for 45 min. at 130 to 140 deg.; sulphuric-acid method is quicker and more accurate than usual nitric-hydrochloric-acid procedure.

**TESTING.** Impact Testing of High Duty Cast Iron. *Foundry Trade Jl.*, vol. 35, no. 546, Feb. 3, 1927, pp. 98-100. Review of paper by J. G. Pearce, and discussion, together with author's reply.

April

## CASTINGS

**DEFECTS IN.** Wasters in the Production of Aluminum-Alloy Castings in Permanent Molds, R. J. Anderson. Foundry Trade J., vol. 35, no. 547 and 548, Feb. 10 and 17, 1927, pp. 122-124 and 151-154, 5 figs. There are four main factors, which have important bearing on occurrence of defects in casting, (1) low density of alloys; (2) their high liquid and low solidification contraction; (3) high coefficient of thermal expansion (in solid state); and (4) lack of strength at elevated temperatures (hot shortness); discusses most important variable factors which govern losses in production of castings by permanent-mold process.

**WATER-COOLED CHILLS.** Water-Cooled Chills. Eng. Progress, vol. 8, no. 2, Feb., 1927, p. 33. Points out disadvantages connected with use of thick-walled cast iron chills when casting nonferrous metals, and employment of thin-walled chills cooled by cooling liquid, which also regulates cooling of fluid metal; describes two types developed in Germany which operate at different predetermined temperatures of cooling liquid such as will influence favorably quality of casting; water-cooled chills have proved specially satisfactory in brass foundries.

## CHROMIUM STEEL

**PROPERTIES.** Facts and Principles Concerning Steel and Heat Treatment, H. B. Knowlton. Am. Soc. Steel Treating—Trans., vol. 11, no. 3, Mar., 1927, pp. 450-463, 4 figs. Influence of chromium upon properties of steel; Guillet's diagram showing structural composition of steels of varying chromium contents; composition, properties, uses and methods of heat treating of number of types of chromium steel; stainless steel and principles of corrosion.

## COPPER

**VISCOSITY.** Viscosity of copper and its Alloys (Essais de viscosité sur le cuivre et ses alliages), J. Cournot and R. Pages. Académie des Sciences—Comptes Rendus, vol. 183, no. 20, Nov. 15, 1926, pp. 885-886. Viscosities have been determined by modification of method previously described at temperatures up to 700 deg.; it has been verified that there exists limiting viscosity and that for slightly greater loads velocity of flow assumes constant value, time required to attain this state being longer the lower the temperature; curve limiting viscosity-temperature for brasses is similar to those for steel; that for copper, however, is approximately rectilinear.

## COPPER ALLOYS

**HARDENABLE.** Notes on the Atomic Behavior of Hardenable Copper Alloys, E. C. Bain. Am. Inst. Min. & Met. Engrs.—Trans., no. 1657-E, Feb. 1927, 8 pp., 3 figs. Results of investigation to discover fundamental atomic conditions existing in Corson's high-copper alloys hardenable by means of silicide solution and reprecipitation; results show that very perfect crystallinity exists in solid solutions prepared at high temperature to contain as much dissolved silicide as possible.

**NICKEL SILVER.** Casting of Nickel Silver, H. Mappin. Metal Industry (Lond.), vol. 30, no. 6, Feb. 11, 1927, p. 164.

Casting consists in heating constituent metals together, and resultant alloy is poured into mold and finally removed as solid ingot; nickel silver, which is alloy of copper, nickel and zinc, is hard bright metal and owing to its high malleability and ductility, it is produced in enormous quantities for manufacture of electroplated articles.

## CORROSION

**THEORIES.** Corrosion in the Gas Industry, I. Ginsberg. Am. Gas J., vol. 126, no. 8, Feb. 19, 1927, pp. 186-188 and 194. General character of rusting; corrosion a form of combustion; electrochemical theory; escape of hydrogen disturbs equilibrium; acid theory; direct chemical action; iron manufacture and corrosion.

## CRUCIBLES

**OVEN FOR.** Describes Ovens for Crucibles, C. F. Hopkins. Foundry, vol. 55, no. 4, Feb. 15, 1927, p. 147, 1 fig. Instances are presented to show how, with proper annealing oven, life of crucibles has been increased from 25 to 50 per cent.

## CRYSTALS

**GROWTH OF.** Growth of Metallic Crystals, C. H. Desch. Min. & Met., vol. 8, no. 243, Mar., 1927, pp. 128-129. First stage of crystallization is appearance of nuclei; second stage is growth of crystals; factors governing rate and character of growth.

**LARGE.** Growing Large Crystals of Metal, W. Darling. Brass World, vol. 23, no. 2, Feb., 1927, pp. 41-42. Difficulties in producing intentionally large crystals of metallic elements and alloys; some huge accidental crystals of iron; their physical properties.

**STRENGTH AND STRAIN HARDNESS.** The Molecular Theory of Strength and Strain Hardness (Zur Molekulartheorie der Festigkeit und der Verfestigung), A. Smekal. Zeit. für technische Physik, vol. 7, no. 11, 1926, pp. 535-544. Investigation to obtain information on heretofore fully unknown molecular state of deformed crystals; it is shown that actual crystals can possess no ideal crystal grid; results are first to provide bases for detailed explanation of slip formation in deformed crystals, which was not possible with all former strain-hardness theories; but all strain-hardness theories are shown to be true, although in widely different degrees.

## CUPOLAS

**FLUORSPAR ADDITIONS.** Rational Use of Fluorspar in Cupolas (Utilisation Rationnelle du Spath Fluor Dans les Cubilots de Fonderie). Fonderie Moderne, vol. 21, Jan., 1927, pp. 13-15. Discusses effects of fluorspar additions.

**FLUORSPAR ADDITIONS.** The action of Fluorspar as Cupola Addition in Iron Foundries (Die Wirkungsweise des Fluorsspath als Kuppelöfen-Zuschlag in der Eisengiesserei), E. Wilke-Dörfurt and T. Klingenstein. Stahl u. Eisen, vol. 47, no. 4, Jan. 27, 1927, pp. 128-133, 7 figs. Application of laboratory desulphurization tests to cupola practice; desulphurization action of limestone and fluorspar; influence of silicic acid of limestone and of cupola lining; change in iron and slag composition; fluorine content of slag.



**LOSSES IN.** Behavior of Cupolas (Sur la conduite des cubilots). *Fonderie Moderne*, vol. 21, Jan., 1927, pp. 10-12, 1 fig. Discusses certain causes of loss and failures of cupolas, citing typical cases in author's own experience.

**PULVERIZED-COAL AUXILIARY FIRING.** Cupolas with Pulverized-Coal Auxiliary Firing (Kuppelofen mit Kohlenstaub-Zusatzfeuerung), U. Lohse. *V. D. I. Zeit.*, vol. 71, no. 7, Feb. 12, 1927, pp. 233-235, 10 figs. Good results have been obtained with use of pulverized coal as additional source of heat in foundry shaft furnace; details of equipment employed.

#### DURALUMIN

**HEAT TREATMENT.** The Heat Treatment of Duralumin, W. Nelson. *Aviation*, vol. 22, no. 8, Feb. 21, 1927, pp. 362-365, 2 figs. When certain light aluminum alloys are heat treated, quenched and aged, there is considerable improvement in their tensile properties; there is leaning towards use of electric furnaces for heat treating this alloy to reduce possible dangers of corrosion from nitrate salts used in baths; this measure appears desirable but uniformity of heating and accurate control of temperature, both obtainable with salt baths, are very important factors.

#### EDUCATION, ENGINEERING

**FOUNDING.** Foundry Technical Institute of France (L'Ecole Supérieure de Fonderie), M. E. Ronceray. *Fonderie Moderne*, vol. 21, Jan., 1927, pp. 1-9. Notes on history, aims, courses and methods of Institute.

#### ELECTRIC FURNACES

**ANNEALING.** Annealing of Metals and Alloys in the Electric Furnace, R. M. Cherry. *Metal Industry (N. Y.)*, vol. 25, no. 2, Feb., 1927, pp. 55-59, 11 figs. Discussion of types of modern electric furnaces for bright and open annealing and their operation.

**DEVELOPMENT.** The Development of Electric Furnaces, C. H. Stevenson. *Fuels and Furnaces*, vol. 5, no. 2, Feb., 1927, pp. 229-230, 3 figs. Design of car-type and tool-room furnaces; firing of chinaware, heat treating of dies; cylindrical furnaces; recent trend.

**DIAGRAM FOR.** A Diagram for Electric Furnaces (Diagramme de fonctionnement des fours électriques), J. Bethenod. *Revue Générale de l'Electricité*, vol. 20, no. 20, Nov. 13, 1926, pp. 697-698, 1 fig. Refers to work by E. de Loisy, published in *Revue de Métallurgie*, May, 1926, developing critical analysis of previous work by P. Bergeon and E. Rieke; describes construction of diagram based upon work of these earlier investigators; by aid of this diagram, functioning of a. c. furnaces can be more closely studied, and results of theoretical investigations can be applied to industrial operation; diagram is of circular type which has been previously used for calculations relating to induction motors; with it, if any one of 6 factors which enter into calculations is known, other 5 can be calculated or deducted from figure. See brief translated abstract in *Science Abstracts (Sec. B)*, vol. 30, part 2, Feb. 25, 1927, pp. 74-75; and reference to article by De Loisy in *Eng. Index*, 1926, p. 257.

**HARDENING.** Salt-Bath Hardening Furnace of German General Electric Co. (Der Salzbadhärteofen der AEG), E. Schmidt. *Centralblatt der Hütten u. Walzwerke*, vol. 30, no. 52, Dec. 29, 1926, pp. 569-570, 5 figs. Details of latest type; under given conditions this furnace can also be used for melting, for instance of white metal, in which case charge is placed in special crucible.

**HEAT TREATING.** Electric Lead Bath Furnace Used in Heat Treating Tools, J. L. Faden. *Fuels & Furnaces*, vol. 5, no. 3, Mar., 1927, p. 366, 1 fig. Furnace used for localized heating of large variety of tools, built by Lord Electric Co., Boston, Mass.

**HEAT TREATING.** Heat Treating of Castings in Electric Furnaces, J. L. Faden. *Fuels & Furnaces*, vol. 5, no. 2, Feb., 1927, pp. 225-226, 3 figs. Castings for machine parts at Whitin Machine Works, Whitinsville, Mass., are carburized and annealed in six box-type electric furnaces which were recently installed.

**HIGH-FREQUENCY.** Contributions to the Theory of High-Frequency Electric Furnaces (Etude théorique du rendement du four électrique à haute fréquence alimenté par alternateur) M. G. Ribaud. *Jl. de Physique et le Radium*, vol. 7, no. 8, Aug., 1926, pp. 250-256, 1 fig. Supplementary to article upon high-frequency furnaces, published in same journal in 1923; mathematical calculations made by Northrup relating to use of alternating currents for high-frequency furnaces are criticized by author as being based upon unreliable hypotheses and results obtained by him are considered untrustworthy; author's examination of theoretical data leads him to conclusion that for each furnace and substance there exists frequency at which output attains maximum and constant figure; this maximum is practically independent of size of furnace, but it decreases as electrical resistance of substance diminishes by heat. See brief translated abstract in *Science Abstracts (Section B)*, vol. 30, part 1, Jan. 25, 1927, p. 21.

**HIGH-FREQUENCY.** The Ironless Induction Furnace (Der eisenlose Induktionsofen), W. Fischer. *Zeit. für technische Physik*, vol. 7, no. 11, 1926, pp. 513-518, 7 figs. Deals with theory, as well as design, connection and operation of high-frequency induction melting furnace for metallurgical practice.

**LABORATORY.** Electric Furnace for Laboratories (Fours électriques pour laboratoires), O. Scarpa. *Chimie & Industrie*, vol. 16, no. 3, Sept., 1926, pp. 219-220, 5 figs. Describes furnace in operation for two years at Electrochemical and Electrometallurgical Laboratory of Royal Polytechnical School at Turin, Italy, which have been of great service in different researches and even in industrial applications; they are similar to Rennerfelt furnace but have the advantage that they can employ both direct and single-phase current.

**MELTING.** Electric Furnaces for Melting Metal. Forging—Stamping—Heat Treating, vol. 12, no. 2, Feb., 1927, pp. 41-43. Limitations and applications of electric furnace; alloy and tool steels; steel castings; cast iron; economics of electric melting; malleable iron; permanent-mold castings; nonferrous metallurgy; electric heating furnaces; electrode control; high-frequency furnace.

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**STEEL.** Electrical Equipment of Arc Steel Furnaces (Ueber die elektrischen Anlagen von Lichtbogen-Elektrostahl-Ofen). K. Kalman. Centralblatt der Hütten u. Walzwerke, vol. 31, nos. 1, 2, Jan. 12, 1927, pp. 3-9, 8 figs. Details of high-tension switches, measuring apparatus, switchboard, furnace transformer, wiring between transformer and furnace; influence of inductive resistances and conductor wiring on economic operation.

**TEMPERING.** Electric Furnace for Tempering Tools. Ry. Elec. Engr., vol. 18, no. 2, Feb., 1927, p. 64, 1 fig. New box or hearth-type electric furnace, particularly applicable to tempering of lathe and planer tools, dies and punches in tool rooms, manufactured by Westinghouse Electric & Manufacturing Co., East Pittsburgh, Pa.

### ELECTRIC WELDING

**STUDIES.** Studies on Electric Welding. L. J. Weber. Am. Soc. for Steel Treating—Trans., vol. 11, no. 3, Mar., 1927, pp. 425-440 and (discussion) 441-449, 6 figs. Study of effect of various gases on metal deposited in electric welding as well as effect of arc on gases; experiments show that abnormal carburizing steels were produced when steel was deposited in atmosphere of  $\text{CO}_2$ ,  $\text{CO}$ , nitrogen and air, while normal carburizing steel was produced when weld was made in atmosphere of helium; analyses of gases show that  $\text{CO}_2$  was decomposed into  $\text{CO}$  and oxygen; oxygen caused oxidation of some iron; nitrogen combines with iron to form nitride, part of which is retained in deposited metal and part of which is condensed on sides of container.

### ELECTRIC WELDING, ARC

**PLATE GIRDERS.** Test of an Arc Welded Plate Girder by the American Bridge Co. and the U. S. Bureau of Standards, H. L. Whittemore. Am. Welding Soc.—Jl., vol. 6, no. 1, Jan., 1927, pp. 42-48, 7 figs. Girder was tested in Olsen compression machine; when tested as simple beam, girder failed by buckling of web plate; none of welds failed until maximum load was reached; increasing size of welds would not have increased maximum load; test showed that large girders of this kind should be carefully designed to use material in most efficient manner.

**PIPE.** Pipe Welding, J. F. Lincoln. Am. Welding Soc.—Jl., vol. 6, no. 1, Jan. 1927, pp. 8-14, 5 figs. Deals with automatic arc welding of pipe and hand welding of field joints for construction of pipe line.

**QUASI-ARC.** Additional Information Supplied by the Quasi-Arc Company Relative to the Australian Tests. Am. Welding Soc.—Jl., vol. 6, no. 1, Jan. 1927, pp. 40-41. Gives diameters of steel cores of quasi-arc electrodes.

**STRUCTURAL STEEL.** Electric Arc Welding of Steel Structures—Design and Organization, H. E. Grove. Am. Welding Soc.—Jl., vol. 6, no. 1, Jan., 1927, pp. 24-40, 16 figs. Outline of procedure adopted in Commonwealth of Australia by superintendent of Metropolitan Gas Co.; in this welding organization no rivets are used, or bolts other than those required for tacking purposes; amount of metal deposited on welded

joint is prescribed by universal meter of 45 deg., gage of electrode used, inches of weld per electrode, and number of runs.

### FOUNDING

**INSTITUTE BRITISH FOUNDRYMEN.** The Early History of the Institute of British Foundrymen, F. J. Cook. Foundry Trade Jl., vol. 35, no. 548, Feb. 17, 1927, pp. 147-149 and (discussion), 149-150. Account of first meeting and first results; initial objects; growth of branches; international connections.

**TECHNICAL TERMS.** A Lexicon of Technical Terms. Foundry, vol. 55, no. 5, Mar. 1, 1927. Foundry data sheet.

### FURNACES

**PATENT SPECIFICATIONS.** Furnaces and Kilns for Applying and Utilizing Heat of Combustion. Abridgements of Specifications, class 51, (ii), period 1916-20 (100,001-155,800), 1926, 396 pp. Patents for inventions.

### FURNACES, ANNEALING

**OIL-BURNING.** Annealing Malleable Iron Castings in an Oil-Burning Furnace, C. C. Mark. Machy. (N. Y.), vol. 33, no. 7, Mar., 1927, pp. 518-520, 1 fig. Locating annealing pots in furnace; packing castings; determining temperature of castings; installation of pyrometers; requirements of furnace; economical operation. See also Machy. (Lond.), vol. 29, no. 745, Jan. 20, 1927, pp. 525-526, 3 figs.

### FURNACES, HEATING

**REGULATION OF ATMOSPHERE.** Furnace Atmosphere is Regulated Automatically by Meters, R. Rimbach and J. A. Stein. Iron Trade Rev., vol. 80, no. 7, Feb. 17, 1927, p. 456. Regulation of furnace combustion conditions to give desired furnace atmosphere is accomplished by use of suitable flue-gas analyzing equipment so as to obtain better uniformity of product with lower fuel consumption; types of furnaces include sheet and bar furnaces, continuous annealing furnaces, heating and reheating furnaces for billets, open and box annealing furnaces and certain forging furnaces; installation often consists of electric  $\text{CO}$  meter, which is sometimes augmented by electric  $\text{CO}_2$  meter.

### FURNACES, INDUSTRIAL

**DESIGN.** Practical Industrial Furnace Design, M. H. Mawhinney. Forging—Stamping—Heat Treating, vol. 12, no. 2, Feb., 1927, pp. 44-47, 5 figs. Various methods of firing furnaces; fuel consumptions and dimensions of interior construction; principles of operation.

### GEARS

**HARDENING.** Hardening Gears with the Oxyacetylene Torch. Acetylene Jl., vol. 28, no. 8, Feb., 1927, pp. 386-388, 1 fig. Describes new method used in England, known as Shorter process, whereby maximum hardness can be imparted to wearing faces of gear-wheel teeth with minimum of distortion; it is claimed that with this process gears can be hardened without distortion.

**HARDNESS**

**TESTING.** Hardness Tester with Impact Tester (Die Härteprüfung mit dem Fallhärteprüfer), V. Schwarz. Motorwagen, vol. 30, no. 3, Jan. 31, 1927, pp. 51-56, 12 figs. Describes tester developed by author by means of which, it is claimed, method of testing is greatly simplified, and determinations can be made quickly, easily and almost without cost.

**IMPACT TESTING**

**NOTCHED-BAR TESTS.** The Law of Similarity in Notched Bar Impact Testing. Metallurgist (Supp. to Engineer), Feb. 25, 1927, pp. 27-29, 2 figs. Moser and others have shown that energy absorbed per unit volume of specimen which suffers permanent distortion under test is approximately independent of size of specimen, and that energy per unit area at fracture is practically constant if specimens are similar but have notches of equal width; in Stahl u. Eisen, Dec. 9, 1926, R. Mailänder discusses validity of Moser's conclusions; it is seen from results that in case of shear failure, there is, at given notch width, an increase of impact strength with size of test piece, whereas for tension failure reverse holds; Moser's statement holds more nearly as tenacity of material tested and size of specimens both become smaller. See reference to original article in A. S. S. T.—Trans., Feb., 1927, p. 312, and Mar., 1927, p. 481.

**IRON**

**FLUORSPAR IN.** Desulphurizing Effect of Fluor Compounds on Molten Iron. Iron & Coal Trades Rev., vol. 114, no. 3077, Feb. 18, 1927, pp. 280-281. Review of experiments conducted by E. Wilke-Dörfurt and H. Buchholz in chemical laboratory of Stuttgart Technical High School, and published in Stahl u. Eisen; results obtained seem to prove fact that readily decomposing fluor compounds are capable of exercising enhanced desulphurizing action on iron and slag.

**OXIDATION.** Chemag Oxidation Process for Iron and Steel. Engineering, vol. 123, no. 3187, Feb. 11, 1927, p. 163. Process invented by A. Mai, of Germany, for oxidation and coloration of iron and steel, consists essentially in immersing articles for periods varying from 5 to 30 minutes, in alkaline bath heated to temperature ranging from 120 to 200 degrees Cent. when uniform tightly adherent film of oxide is deposited on surface; invention depends on discovery that presence of caustic alkaline earth exerts favorable influence on oxidation process.

**RUSTLESS.** Making Rustless Iron, D. St. Pierre du Bose. Iron Trade Rev., vol. 80, no. 7, Feb. 17, 1927, pp. 457-459, 2 figs. Recently developed alloy of iron and chromium gives wide range of physical properties when tempered or annealed.

**STAINLESS.** Stainless Iron, C. B. Callomon. Oil. Bul., vol. 13, no. 2, Feb., 1927, pp. 169-173. In annealed condition stainless iron has tensile strength of 75,000 lb. per sq. in. with elastic limit of 45,000 lb. per sq. in., elongation of 35 to 40 per cent, and reduction in area of from 75 to 80 per cent; what it will do in combating corrosion.

**IRON ALLOYS**

**IRON-CHROMIUM NICKEL.** An Introduction to the Iron-Chromium-Nickel Alloys, E. C. Bain and W. E. Griffiths. Am. Inst. Min. & Met. Engrs.—Trans., no. 1650-C, Mar., 1927, 46 pp., 37 figs. Results of inquiry into structural nature of some 70 iron alloys containing both nickel and chromium over considerable range of concentration; above 1000 degrees Cent. entire iron-chromium-nickel system is probably composed of only two kinds of solid solutions; one is austenite containing some chromium and other is ferrite containing some nickel; ferrite phase increases in composition range with rise in temperature above 1050 degrees Cent.; many of alloys containing from 5 to 10 per cent each of nickel and chromium are martensitic when quenched from great variety of temperatures and at great variety of rates.

**IRON AND STEEL**

**CONSTITUTION.** The Constitution of Iron and Steel, H. M. Boylston. Fuels & Furnaces, vol. 5, no. 3, Mar., 1927, pp. 297-304, 6 figs. With special reference to heat treatment; determination of critical points; effect of carbon content; influence of rate of heating and cooling; importance of proper temperature measurements.

**IRON CASTINGS**

**OIL-SAND APPLICATION.** Oil Sand and Motor Castings, W. West and W. Aston. Foundry Trade J., vol. 35, no. 547, Feb. 10, 1927, pp. 125-130, 12 figs. Methods employed for manufacture of cylinders and other castings in use at Leyland Works.

**IRON ORE**

**DIRECT REDUCTION.** Direct Reduction of Iron Ore. Iron Age, vol. 119, no. 7, Feb. 17, 1927, pp. 498-499, 2 figs. Dutch method, invented by D. Croese, avoids blast furnace; electricity used in reducing iron sands; crushed ores to be handled.

**DIRECT REDUCTION.** Eliminating Blast Furnace in Norway. Iron Age, vol. 119, no. 8, Feb. 24, 1927, pp. 563-565, 7 figs. Process of gas reduction of iron ore, as solved by A. S. Norsk Staal, involves system of complete mechanical automatic operation; no direct handling is required, material being automatically brought from storage and treated, and briquetted iron sponge delivered, without any manual assistance whatsoever.

**MANGANIFEROUS.** Utilization of Manganiferous Iron Ores, T. L. Joseph, P. H. Royster and S. P. Kinney. U. S. Bureau of Mines—Tech. Paper, no. 393, 1926, 28 pp., 17 figs. Investigation on use of manganiferous iron ores with respect to effect on blast-furnace operation; effect of high-manganese pig iron on quality of steel, decrease in ferromanganese by use of high-manganese pig iron, and desulphurization of steel through presence of manganese; utilization of manganiferous iron ores in production of ferromanganese; use of 6 to 10 per cent manganese alloys as additions to steels containing more than 0.3 per cent carbon.

**IRON, PIG**

**CARBON IN.** Carbon in Pig Iron. Min. & Met., vol. 8, no. 243, Mar., 1927, pp.

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140-141. Review of round-table conference. Need for research in foundry pig iron; carbon characteristics of copper-bearing pig; amount of carbon pig iron can contain; use of charcoal and coke in production of pig.

**FOUNDRY.** Need for Research in Foundry Pig-Iron. R. Moldenke. Am. Inst. Min. & Met. Engrs.—Trans., no. 1648-C, for mtg. Feb., 1927, 5 pp. Conditions under which pig iron is melted; trend today is toward specification of total carbon content in pig irons; it seems that time has come for technical side of furnace development of country to be organized into associated effort, so that this industry may benefit in similar manner as has that of foundry.

### LEAD ALLOYS

**LEAD-ANTIMONY.** The Solidus Line in the Lead-Antimony System. E. E. Schumacher and F. C. Nix. Am. Inst. Min. & Met. Engrs.—Trans., no. 1636-E, Feb., 1927, 11 pp., 13 figs. Investigation of solidus line above solid solution field for lead-antimony system by quenching-test procedure has enabled writers to determine 3 points between melting point of pure lead and end of eutectic horizontal, and therefore to fix precise location of this line.

### MAGNESIUM ALLOYS

**ANALYSIS.** Methods of Analyzing Industrial Alloys of Magnesium (Methodes d'analyse des alliages industriels de magnesium), E. Pretet and L. Ecoffet. Chimie & Industrie, vol. 16, no. 3, Sept., 1926, pp. 459-470, 2 figs. Determination of silicon, iron, aluminum, copper, zinc, manganese, lead, cadmium, nickel, calcium, carbon, nitrogen, etc., in magnesium and magnesium alloys.

### MALLEABLE CASTINGS

**PRODUCTION 1926.** Production of Malleable Castings 1,000,000 Tons. Iron Age, vol. 119, no. 9, Mar. 3, 1927, p. 634. Output in 1926 about on par with two preceding years.

### MANGANESE STEEL

**PROPERTIES.** Curious and Hitherto Unexplained Facts Concerning Manganese Steel (Quelques faits curieux et non encore expliqués observés dans l'acier au manganèse), R. Hadfield. Chimie & Industrie, vol. 16, no. 3, Sept., 1926, pp. 136-139, 4 figs. Properties of manganese steel; effects of heat treatment on hardness and magnetic properties; photomicrographic examination.

### METAL WORKING

**PATENT SPECIFICATIONS.** Metal Articles and Forms, Combination Apparatus and Processes Specially Designed for Producing and Treating. Abridgements of Specifications, class 83 (ii), period 1916-20, 1927, 186 pp. Patents for inventions.

### METALLOGRAPHY

**MICROSCOPIC EXAMINATION OF METALS.** The Technique of Examining Metals under the Microscope. Gen. Elec. Rev., vol. 30, no. 3, Mar., 1927, pp. 160-166, 16 figs. Microscopic and photographic optics.

### METALS

**CORROSION.** Controllable Variables in the Quantitative Study of the Submerged Corrosion of Metals, O. B. J. Fraser, D. E. Ackerman and J. W. Sands. Indus. & Eng. Chem., vol. 19, no. 3, Mar., 1927, pp. 332-338, 9 figs. Great influence of controllable variables on results of laboratory corrosion tests of total-immersion type is discussed and illustrated by numerous data obtained in study of action of sulphuric acid solutions on monel metal.

**CRYSTALLITE ORIENTATION.** Crystallite Orientation in Metallic Workpieces in its Relation to the Elastic Properties (Die Kristallitenorientierung in metallischen Werkstücken in ihrer Beziehung zu den elastischen Eigenschaften), G. Tammann. Zeit. für technische Physik, vol. 7, no. 11, 1926, pp. 531-534. Points out that compressive strength and other elastic properties are not alone dependent on nature of material, but also on orientation of crystallites of which material is composed; discusses X-ray examination methods, methods of maximum reflection, and etching.

**DEFORMATION.** Hot and Cold Deformation of Metals (Ueber Warm- und Kaltverformung der Metalle), F. Sauerwald and H. Giersberg. Centralblatt der Hütten u. Walzwerke, vol. 30, nos. 47 and 49, Nov. 24 and Dec. 8, 1926, pp. 501-504 and 525-529, 13 figs. Investigation of spontaneous crystallization in connection with deformation at higher temperatures, relation of temperature to age hardening, with special consideration of time factor; results aid in establishing with greater accuracy relation between hot and cold deformation; crystallization and its relation to deformation temperature and time.

**DRILLING TOOLS.** Hard Metallic Materials for Drilling Tools. Petroleum Times, vol. 17, no. 422, Feb. 12, 1927, p. 310. According to information supplied by A. Merz and W. Schulz, published in Glückauf, there have been discovered eight different metallic materials which possess properties of special value for use as cutting parts of drilling tools; besides these, there are alloys known as Volumit, Thoran, Stellite, Hartmetall, Celsit, Akrit; these different materials give range suitable for any type of drilling from rocks of moderate hardness up to hardest rocks known.

**ELASTIC PROPERTIES.** Study of Elastic Properties and Viscosity of Metals and Alloys (Contribution a l'étude des propriétés élastiques et de la viscosité des métaux et alliages), P. Chevenard and A. Portevin. Chimie & Industrie, vol. 16, no. 3, Sept., 1926, pp. 434-448, 35 figs. Methods and apparatus for study of elasticity; modulus of elasticity and internal friction as function of initial state and of testing temperature; influence of quenching; study of viscosity of alloys.

**ENDURANCE.** The Endurance of Materials, its Practical Importance and its Determination by Means of New Types of Testing Machines (Die Dauerfestigkeit, ihre Bedeutung für die Praxis und ihre kurzfristige Ermittlung mittels neuartiger Prüfmaschinen), E. Lehr. Glasers Annalen, vol. 50, nos. 8, 9, 12, Oct. 15, Nov. 1, Dec. 15, 1926, pp. 109-114, 117-122, 177-180, and vol. 50, no. 3, Feb. 1, 1927, pp. 33-39,

26 figs. Examples are given of light engines, bridges, springs, ships, etc., to illustrate importance of endurance properties of materials; characteristics of fatigue phenomena; new methods and machines for observation of physical changes occurring at limit of fatigue, which permits determination of endurance properties of materials by means of rapid tests; machines have been developed thus far for tensile stress, bending and torsion; gives numerous results of tests.

**FATIGUE.** The Fatigue of Metals. Machy. (Lond.), vol. 29, no. 747, Feb. 3, 1927, pp. 596-597, 3 figs. Failures due to fatigue; effect of surface cracks.

**FUTURE DEMAND FOR.** Future Demand for Metals, H. F. Bain. Am. Inst. Min. & Met. Engrs.—Trans., no. 1601-I, Oct., 1926, 12 pp. Increasing dependence on minerals and larger individual consumption of them is most striking in United States; deals with question of reserves which, it is claimed, rests upon technology and economics; it is a matter of developing methods of finding and working such ores at a cost the world can pay; two trends are discernible in modern production, which are accumulation of stocks in use of metals reclaimable after use, and ability to make substitutions.

**HARDNESS.** Theory of Hardness of Metals (Sur la théorie de la dureté des métaux), K. Honda. Chimie & Industrie, vol. 16, no. 3, Sept., 1926, pp. 420-426, 5 figs. Discusses five different theories of hardness of metal, namely, (1) theory of beta iron, (2) amorphous state, (3) hot solution, (4) interstrain, and (5) slip interference.

**PLASTICITY AND RECRYSTALLIZATION.** Plasticity, Strain Hardness and Recrystallization (Ueber Plastizität, Verfestigung und Rekristallisation), R. Becker. Zeit. für technische Physik, vol. 7, no. 11, 1926, pp. 547-555, 2 figs. Structure of deformed material is conceived as intermediate stadium between single crystal and fully amorphous subcooled liquid; this interpretation offers explanation of recrystallization diagram, of change of density with deformation, and of occurrence of place changing and consequent "amorphous plasticity".

**TENSILE TESTS.** Tensile Tests of Flat Tensile Bars (Der Zugversuch am Flachstab), W. Kuntze and G. Sachs. Stahl u. Eisen, vol. 47, no. 6, Feb. 10, 1927, pp. 219-226, 14 figs. Influence of expansion and shrinkage; relation of cross-section on stress and deformation; expansion and shrinkage in copper or similar metals, and in iron.

**THERMAL EXPANSION.** Influence of Working on Thermal Expansion Coefficient of Metals (Einfluss der Bearbeitung auf den thermischen Ausdehnungskoeffizienten der Metalle), W. Jubitz. Zeit. für technische Physik, vol. 7, no. 11, 1926, pp. 522-527, 9 figs. Influence of plastic working, such as rolling, forging, drawing, etc., is investigated experimentally on number of metals; for metals belonging to regular system (bronze, low-carbon steel), no influence on expansion coefficients could be established; for metals that do not belong to regular crystal system (zinc, cadmium, magnesium), such influence was observed.

#### MOLDING METHODS

**HOLLOW CYLINDERS.** How to Make a Mold for a Long Hollow Cylinder. Foundry,

vol. 55, no. 5, Mar. 1, 1927, pp. 174-176, 2 figs. Several alternative methods suggested; local conditions affect final choice; loam and dry sand compared as molding medium.

#### NICKEL

**STRUCTURE.** The Structure of Nickel, H. Collins. Chem. News, vol. 134, no. 3487, Feb. 11, 1927, pp. 81-85. Constitution of an atom of nickel has previously been given as SIP; this will now be proved to be correct quite independently by means of laws of relative volume and of heat of formation.

#### NICKEL STEEL

**ELECTRICAL RESISTANCE.** Dependence of Electrical Resistance of Nickel Steel on Composition, Temperature, and Heat Treatment (Ueber die Abhängigkeit des elektrischen Widerstandes der Nickelstähle von Zusammensetzung, Temperatur und Wärmebehandlung), F. Ribbeck. Zeit. für Physik, vol. 39, no. 10-11, Nov. 16, 1926, pp. 787-812, 11 figs. Curves and tables are given showing variation of resistance with temperature from -180 to 1000 degrees Cent. for steels with varying nickel content; irreversibility of steels with less than 30 per cent Ni is attributed to supercooling.

#### OPEN-HEARTH FURNACES

**COOLING.** Cooling of Open-Hearth Furnaces (Die Kühlung von Siemens-Martin Oefen), G. Bulle. Stahl u. Eisen, vol. 47, nos. 2 and 3, Jan. 13 and 20, 1927, pp. 41-52, and 85-90, 37 figs. Purpose of cooling; different arrangements and devices; heat requirement, durability and operating costs; advantages and disadvantages of cooling.

**DESIGN.** Open-Hearth Furnace Construction, C. W. Veach. Blast Furnace & Steel Plant, vol. 15, no. 2, Feb., 1927, pp. 92-93. It is desirable to know fusing point of various refractories entering into furnace construction; kind and quality of brick stressed; chromium ore used.

**DEVELOPMENTS.** Developments in the Open-Hearth Process, B. M. Larsen. Forging—Stamping—Heat Treating, vol. 12, no. 2, Feb., 1927, pp. 55-60. Review of furnace construction and operation; metallurgical practices; temperature measurement and control; waste-heat boilers; insulation; refractories; pig iron for open-hearth; gas oxidation. Bibliography.

**EXHAUST-GAS UTILIZATION.** Recovery of Flue Dust Containing Zinc and Lead Oxide, from Exhaust Gas of Open-Hearth Furnaces (Die Gewinnung von zinn- und bleioxydhaltigem Flugstaub aus den Abgasen eines Siemens-Martin-Ofens), S. Schleicher. Stahl u. Eisen, vol. 47, no. 5, Feb. 3, 1927, pp. 169-172, 4 figs. Quantity and composition of flue dust contained in exhaust gases; measurement of exhaust gases; description of Cottrell-Möller dust-precipitation plant and operating results.

**FUEL-GAS PURIFICATION.** The Purification of Fuel Gas for Open-Hearth Furnaces, F. W. Sperr, Jr. Fuels and Furnaces, vol. 5, no. 3, Mar., 1927, pp. 363-365, 2 figs. Liquid purification process for purification of gas for open-hearth furnaces; sulphur-recovery process for purification of manufactured gas.

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**OXYACETYLENE WELDING.** Oxygen and Acetylene About the Open-Hearth Furnace. Oxy-acetylene Tips, vol. 5, no. 7, Feb., 1927, pp. 132-141, 22 figs. Points out that oxy-acetylene welding has proved itself a material factor in construction maintenance, operation and dismantling; furnace tapping with oxygen lance; relining furnaces; pouring hot sheet; slag-ladle repairs.

**STEEL-MELTING.** Small Open-Hearth Furnace of New Design, H. G. Begeman. Fuels & Furnaces, vol. 5, no. 2, Feb. 1927, p. 242, 2 figs. Furnace for melting steel, with gas producers built on to both heads of furnace, has recently been developed and placed in operation in steel plant in Austria; hearth of furnace is connected with high-temperature zone of producers, so that radiant heat of this zone acts directly upon hearth space and increases its temperature.

**WATER COOLING.** Water Cooling of Open-Hearth Furnace Heads (Wasserkühlung an Siemens-Martin-Ofenköpfen), A. Ziegler. Stahl u. Eisen, vol. 47, no. 3, Jan. 20, 1927, pp. 98-99, 3 figs. Describes cooling system developed for 25-ton furnace at steel works in Röchling, Germany, and good experiences obtained therewith.

## OXYACETYLENE WELDING

**AIRPLANE CONSTRUCTION.** Autogenous Welding in Airplane Construction (Die autogene Schweissung im Flugzeugbau), L. Kuchel. Autogene Metallbearbeitung, vol. 20, no. 1, Jan. 1, 1927, pp. 3-9, 9 figs. Application to welding of steel tubes and other construction parts.

**TORCH MANIPULATION.** Torch Manipulation in Welding. Welding Engr., vol. 12, no. 2, Feb., 1927, pp. 30-33, 21 figs. Movements of torch are not just to keep it on jump; every motion of torch has effect on weld.

## REFRACTORIES

**SLAG AND FLUE DUST, EFFECT OF.** Testing Method for Quantitative Determination of Attack of Slags and Flue Dust on Refractory Brick (Ein Prüfverfahren zur quantitativen Bestimmung des Angriffs von Schlacke und Flugstaub auf feuerfeste Steine), F. Hartmann. Stahl u. Eisen, vol. 47, no. 5, Feb. 3, 1927, pp. 182-186, 3 figs. Weight and volumetric loss as a measure of absorptivity; influence of temperature, porosity, etc.; slagging effect of iron and manganese oxide; application to testing of firebrick and raw materials.

## RESEARCH

**METALS.** Industrial Metal Research (Metallforschung in der Industrie), E. H. Schulz. V. D. I. Zeit., vol. 71, no. 6, Feb. 5, 1927, pp. 185-188, 2 figs. Review of development and special position of metal research in industry; relations between industrial research and research in pure science; future prospects.

## ROLLING MILLS

**BEAMS.** Material Flow in the Rolling of Beams, N. Metz. Blast Furnace & Steel Plant, vol. 15, no. 2, Feb., 1927, pp. 82-87, 38 figs. In order to determine flow of metal when being rolled, method has been devised in which threaded test pieces are imbedded in block of steel and this

block rolled exactly as in practice. Translated from Stahl u. Eisen, Nov. 16, 1927, pp. 1577-1582.

**COLD STRIP.** Coilers for Cold Strip Rolling Mills, C. E. Davies. Engineer, vol. 143, nos. 3707 and 3708, Jan. 28 and Feb. 4, 1927, pp. 92-94, and 123-125, 21 figs. Primary purpose is merely to dispose conveniently of long strips of metal; hand coiling is only advantageous when dealing with very short strips; types of coilers; essential features of effective coiler of regular type; principle of automatic coiler; thick strip coilers.

**ELECTRIC DRIVE.** Electrical Developments. Blast Furnace & Steel Plant, vol. 15, no. 2, Feb., 1927, pp. 80-81, 2 figs. Increase in number of motors is indication of trend toward individual drive for various stands; tendency toward large prime movers.

**FORD MOTOR CO., RIVER ROUGE.** The Steel Mill of the Ford Motor Company, T. Harvey. Gen. Elec. Rev., vol. 30, no. 3, Mar., 1927, pp. 152-159, 5 figs. Rolling mill; system of power generation and distribution; merchant and blooming-mill motors and control; projected developments and equipment.

## SCRAP

**CLASSIFICATION.** Classification of Old Metals. Brass World, vol. 23, no. 2, Feb., 1927, pp. 55-56. Presents official terms in scrap-metal trade for reference purposes.

## STEEL

**ANALYSIS.** Some Notes on the Analysis of Steels, W. Singleton. Chem. Age (Met. Section), vol. 16, no. 397, Feb. 5, 1927, pp. 9-11. Methods of analysis for determination of chromium, manganese, molybdenum and small amounts of copper when present in steel; they consist for the most part of modifications of existing methods and are of particular interest because of their rapidity and ease of operation.

**AUSTENITE TRANSFORMATION INTO MARTENSITE.** On the Transformation of Retained Austenite into Martensite by Stress, K. Honda and K. Iwase. Am. Soc. for Steel Treating—Trans., vol. 11, no. 3, Mar., 1927, pp. 399-406 and (discussion) 407-412, and 473-474, 6 figs. Quenched austenite can be transformed into martensite by cold working; internal stress set up in steel by quenching promotes transformation of retained austenite into martensite; fact observed by Mathews that in case of alloy steels, more austenite is obtained by oil than by water quenching, is only valid for thick plates or rods in which internal stress caused by rapidly cooling is fairly large, and can be explained as effect of stress; by cooling quenched steels in liquid air, retained austenite is transformed into martensite, amount of transformation being less the thicker the specimens.

**AUSTENITIC STRUCTURE.** The Decomposition of the Austenitic Structure in Steels, O. E. Harder and R. L. Dowdell. Am. Soc. for Steel Treating—Trans., vol. 11, no. 3, Mar., 1927, pp. 391-397, 12 figs. Continuation of research on decomposition of austenitic structure in steel, paying particular attention to decomposition of austenite in liquid oxygen; six steels were included in investigation; (1) cobalt-chromium mag-

net steel, (2) Hadfield manganese steel, (3) high-carbon high-chromium steel, (4) high-speed steel, (5) 22-per cent nickel steel, and (6) hypereutectoid carbon steel; marked difference in stability of austenitic structure at low temperatures is clearly shown; consideration of influence of stresses at liquid-oxygen temperature on decomposition of austenite.

**BRITTLENESS.** Influence of Compression on the Brittleness of Steel (Influence de la compression sur la fragilité de l'acier), P. Dejean. Académie des Sciences—Comptes Rendus, vol. 184, no. 4, Jan. 1927, pp. 188-189. Results of number of tests on bars with square section of 32 mm. on side forged from same billet and with same carbon content (0.80); these bars were divided into three groups and heat treated in different manners; from results it appears that as long as load on bar does not attain certain critical value, resilience of metal does not seem to be materially affected by previous compression, but beginning with that critical value resilience falls off very suddenly; this critical value varies with previous treatment of bar.

**CEMENTATION.** Cementation of Mild Steel (Cementation de l'acier doux par le cyanogène et la cyanamide), E. Perot. Académie des Sciences—Comptes Rendus, vol. 183, no. 23, Dec. 6, 1926, pp. 1108-1110, 2 figs. Use of cyanogen and of cyanamide for cementite formation as compared with ethylene and methane; cyanogen without renewing atmosphere gives results about equivalent to those obtained with ethylene or methane in continuous circulation.

**CEMENTITE CONDITION.** The Importance of Cementite, R. G. Guthrie. Am. Soc. for Steel Treating—Trans., vol. 11, no. 3, Mar., 1927, pp. 341-354, 17 figs. Author calls special attention to what he believes to be basic and well-known points of vital importance when anticipating behavior of articles manufactured from straight carbon steel; he is of opinion that condition of cementite is of utmost importance, and that in low-carbon steels which are not to be subjected to hardening, it is desirable that cementite be in spheroidal condition but that in high-carbon steels which are to be subjected to hardening it is desirable to have cementite in lamellar form.

**COLD-DRAWN.** Volumetric Changes in Steel with Cold-Drawing (Volumenänderungen von Stahl beim Kaltrecken), E. Houdremont and E. Bürklin. Stahl u. Eisen, vol. 47, no. 3, Jan. 20, 1927, pp. 90-93, 3 figs. Importance of volumetric change in Mauser hardness theory; volume change in hardened and cold-drawn steels in relation to degree of deformation; points out that there is no absolute relationship between increase in strength and decrease in density; stress distribution with hardening and cold drawing; results of extended drawing tests.

**CORROSION FATIGUE.** Corrosion-Fatigue of Metals as Affected by Chemical Composition, Heat Treatment and Cold Working, D. J. McAdam, Jr. Am. Soc. for Steel Treating—Trans., vol. 11, no. 3, Mar., 1927, pp. 355-380 and (discussion) 380-390, 12 figs. Gives results of previous investigation of corrosion fatigue of metals and sets forth materials, machines and specimens used in further investigation; author describes and illustrates corrosion-fatigue spots and is of

opinion that transverse cracks passing through non-metallic inclusions usually surrounded by oxide coating are origins of corrosion-fatigue tests; tests were made in series of carbon steels ranging from 0.033 to 1.09 per cent carbon; experiments with carbon steels, nickel steels, high-chromium steels and chromium-nickel steels; in author's opinion corrosion fatigue depends on two factors, corrosion intensity and stress range, and corrosion-fatigue limit depends on strength factor as well as on corrosion resistance.

**EMBRITTEMENT.** Embrittlement of Steel, A. G. Christie. Am. Water Works Assn.—Jl., vol. 17, no. 2, Feb., 1927, pp. 174-189. Progress report of Subcommittee No. 6 on embrittlement of metals; certain engineers have been led to conclusion that such embrittlement as occurs is promoted by presence of caustic in waters in boiler resulting from occurrence of sodium carbonate in feedwater; other engineers have concluded that other causes contribute to this type of failure and that presence of caustic is not controlling factor; evidence and reasoning of both schools of thought are summarized; most experimental work carried on has been with waters high in carbonate or bicarbonate of soda.

**FATIGUE.** Hysteresis and Fatigue of the Wöhler Rotating Cantilever Specimen, N. P. Inglis. Metallurgist (Supplement to Engineer), Feb. 25, 1927, pp. 23-26, 6 figs. Experiments definitely indicate existence of fatigue limit, but not as definite point on work curve; so far as experiments have gone, there appears to be no simple relation between work done up to fracture at given stress and that stress; tests were carried out on bars made from batches of low-carbon and medium-carbon steel.

**FATIGUE STRENGTH.** Fatigue Strength of Hard Steels and their Relation to Tensile Strength, J. M. Lessells. Am. Soc. Steel Treating—Trans., vol. 11, no. 3, Mar., 1927, pp. 413-1420 and (discussion) 1421-1424, 10 figs. Data on tensile and fatigue properties of very hard steels; application of the extensometers to measurement of elastic properties; relationship between tensile strength, Brinell hardness and endurance limit; shows how residual stresses present in material may influence such relationship; superiority of medium-carbon steel in gear application is shown by results obtained.

**IDEAL VS. COMMERCIAL.** Ideal Versus Commercial Steels, J. D. Gat. Forging—Stamping—Heat Treating, vol. 12, no. 2, Feb., 1927, pp. 38-40 and 43, 2 figs. Composition of steel and effect of alloys discussed from metallurgical viewpoint; influence of inclusions.

**MOLYBDENUM IN.** Molybdenum in Steel, Iron & Coal Trades Rev., vol. 114, no. 3077, Feb. 18, 1927, p. 279. Review of report issued by Research Department Woolwich, on influence of molybdenum on medium-carbon steels containing nickel and chromium.

**NICKEL AND COBALT IN.** A Comparison of the Effect of Nickel and Cobalt in Steel, F. H. Allison, Jr. Am. Inst. Min. & Met. Engrs.—Trans., no. 1662-C, Mar., 1927, 11 pp., 12 figs. Chemically, both elements enter into steel in much the same manner; they are found to dissolve freely in ferrite, and are found also in solution with iron in

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carbide; both render carbide unstable, and nickel is probably the more deleterious agent; differences between nickel and cobalt steels are caused by individual effects which these elements exert on critical points.

**NITRATION.** Nitration of Steels (Sur la nitruration des aciers), L. Guillet. Académie des Sciences—Comptes Rendus, vol. 183, no. 21, Nov. 22, 1926, pp. 933-935. Brinell tests have been carried out on tempered case-hardened steel, and on chrome-aluminum steel which had been nitrated by means of ammonia for 90 hr. at 510 deg. so as to produce nitrated layer 0.8 mm. thick; latter steel had greater initial hardness, and this was retained to greater extent than in case of former, when steels were maintained at gradually increasing temperatures (from -180 to 600 deg.) for various periods of time.

**SEGREGATION.** Segregation and Strength Properties (Seigerungen und Festigkeitseigenschaften), G. Fiek and G. Sachs. Bauingenieur, vol. 8, no. 5, Jan. 29, 1927, pp. 75-78, 6 figs. Results of investigations made by German State Materials Testing Institute on I-beams, showing deleterious effect of phosphorus-rich segregations in low-carbon steels; deformability of segregated parts is considerably less than that of pure material.

**SHEET, GALVANIZED.** A Theory of the Cause of Blisters on Galvanized Sheets, L. B. Lindemuth. Am. Inst. Min. & Met. Engrs.—Trans., no. 1659-C, Mar., 1927, 3 pp. Author's theory is that blisters are caused by moisture that has penetrated steel after annealing operation; this moisture comes from atmosphere, or from pickling or washing, and when it is suddenly expanded into superheated steam in galvanizing, increase in volume is so great that it cannot escape with sufficient rapidity through small opening where it entered as moisture; prevention of blisters.

**STRESS DISTRIBUTION.** Stress Distribution in Mild Steel as Indicated by Special Etching, J. D. Jevons. Engineering, vol. 123, nos. 3187 and 3189, Feb. 11 and 25, 1927, pp. 155-157 and 221-223, 41 figs. partly on supp. plate. Investigation of strain-etch figures produced on specimens of various shapes; standard cylinders under compression; notched tensile bars; influence on tensile specimens of Dalby collar; effect of direction of rolling on strain figures produced on rolled bars; influence on strain figures of chemical composition and type of steel; mechanism of deformation of mild steel; comparison of strain-etch figures with figures obtained by photoelastic methods and by scale on surface effects.

**TEMPER BRITTLENESS.** Temper Brittleness. Metallurgist (Supplement to Engineer), Feb. 25, 1927, pp. 21-23, 2 figs. For steels which show temper brittleness there is temperature range in which, starting either from tough or brittle condition, tempering produces intermediate notched-bar impact value corresponding to equilibrium condition; tempering above this range always produces tough condition and below it causes no change; refers to paper by Guillet and Ballay in Revue de Métallurgie, giving results of repeated impact tests indicating that specimens showing temper brittleness have lower values, especially if energy of repeated blows is increased; tests were also

made at increasing temperatures on tough and embrittled specimens of nickel-chromium steels; includes list of papers presenting interesting theories to explain temper brittleness.

## STEEL CASTINGS

**GAS SOLUBILITY IN.** Solubility of Gases in Cast Steel (Recherches sur la solubilité des gaz dans l'acier fondu), P. Dejean. Chimie & Industrie, vol. 16, no. 3, Sept., 1926, pp. 427-429, 3 figs. Results of investigation to determine influence of nature of gases on formation of blowholes.

**RISERS, REMOVAL OF.** The Removal of Risers in the Steel Foundry, L. E. Everett. Am. Welding Soc.—Jl., vol. 6, no. 1, Jan., 1927, pp. 48-56. Explains methods of removing risers used in modern progressive steel foundries; discusses four methods of removing heads: (1) by flogging or slogging, (2) by sprue cutter, (3) by cold saw (high or low-speed), (4) by oxyacetylene cutting torch; deals particularly with last method.

## STEEL, HEAT TREATMENT OF

**DRILL STEEL.** Carbon Drill Steel and its Heat Treatment, V. O. Cutts. Min. & Indus. Mag., vol. 3, no. 11, Jan. 26, 1927, pp. 383-388. Steel failures and their correction; care in selecting steel; comparison between heat treated and normalized steels; getting maximum and best service; standard test for drill steels.

**HARDENING.** Influence of Temperature of Quenching on Mechanical Properties of Special Low-Carbon Steels (Influence de la température de trempe sur les propriétés mécaniques d'aciers spéciaux peu carbures), M. Sauvageoy. Chimie & Industrie, vol. 16, no. 3, Sept., 1926, pp. 411-419. Results of tempering tests; micrographic examination of specimens; determination of brittleness of bars after quenching and heating at different temperatures and determination of hardness of specimens taken from these bars; results of tensile tests of bars under different treatments.

**HARDENING.** Pack Hardening of Various Steels, A. Mumper. Forging—Stamping—Heat Treating, vol. 12, no. 2, Feb., 1927, pp. 53-55 and 60, 4 figs. Mention is made of several grades of steel used in making dies; heat treatment for various parts of dies; pack-hardening of tool steel often advantageous.

**LABORATORY FINDINGS APPLIED TO.** Laboratory Aids Steel Treater, A. W. F. Green. Iron Age, vol. 119, no. 7, Feb. 17, 1927, pp. 487-490, 11 figs. Specific instances of value of co-operation; relation of laboratory operations to steel treating.

**TIRES AND AXLES.** The Forging and Heat Treatment of Tyres and Axles, W. Sutherland. Australasian Inst. Min. & Met.—Proc., no. 63, Sept., 1926, pp. 33-46, 1 fig. Iron-iron carbide equilibrium diagram; considers changes which occur on cooling pure carbon steels of varying carbon content from temperatures above upper limit of critical range, at which steel consists of solid solution referred to as austenite; deals with annealing, normalizing, hardening and tempering; mechanical treatment of steel; manufacture and testing of tires and axles.



**TOOL STEEL.** Tool Hardening and Annealing in Railway Repair Shop at Schwerte (Die Werkzeughärterei und Glüherei im Eisenbahnausbesserwerk Schwerte), L. Kupfer and F. Böhm. Archiv für Wärmewirtschaft, vol. 8, no. 1, Jan., 1927, pp. 9-12, 11 figs. Suggestions for use of gas; measures for obtaining uniform temperatures in annealing room; results of tests on annealing and hardening furnaces; elimination of disturbing influences in pipe line.

### STEEL MANUFACTURE

**INGOT PRODUCTION.** Production of Steel Ingots, H. M. Boylston. Fuels & Furnaces, vol. 5, no. 2, Feb., 1927, pp. 163-170, 8 figs. Teeming of ingots; teeming ladle; temperature and rate of pouring various types of ingot molds; stripping of ingots; soaking pits, their construction and use.

**METHODS.** The Qualities of Steel and their Relations to Manufacturing Methods (Ueber Stahlqualitäten und ihre Beziehungen zu den Herstellverfahren), P. Goerens. Krupp'sche Monatshefte, vol. 8, Jan., 1927, pp. 1-8, 10 figs. What is meant by quality of steel; nonmetallic inclusions in steel, and their significance; advantages and disadvantages of present methods of manufacturing steel; points out necessity for close co-operation between steel users and manufacturers.

### STEEL WORKS

**FUEL ECONOMY IN.** Fuel Economy in Steel Works Producing its own Coke (L'utilisation des combustibles dans une usine sidérurgique complète), M. Laffargue. Chimie & Industrie, vol. 16, no. 3, Sept., 1926, pp. 391-398. Points out that steel plant having its own coke-oven plant has considerable amount of excess gas, and complete utilization of this combustible gas is seldom realized; author recommends organization of heat-control department, which would effect economy in fuel far surpassing cost of maintaining such a department.

**POWER INSTALLATIONS.** Use of Power in American Steel Mills, W. Sykes. Iron Age, vol. 119, no. 9, Mar. 3, 1927, p. 635. Paper read before Mid-West Power Conference. (Abstract.)

**SPAIN.** Integrated Steel Mill in Spain, F. C. Roberts. Iron Age, vol. 119, nos. 7 and 8, Feb. 17 and 24, 1927, pp. 494-497 and 571-573, 5 figs. Feb. 17: Ore from nearby mountains joins imported coal at seaboard; pig iron, billets and finished lines made; located at port of Sagunto, Spain, on shore of Mediterranean Sea. Feb. 24: Reversing blooming mill, three-high rail and structural mill, plate and merchant mills work up ingot output. See also description in Iron & Coal Trades Rev., vol. 114, no. 3076, Feb. 11, 1927, pp. 218-220, 2 figs.

### STELLITE

**CHARACTERISTICS.** Stellite, A. V. Harris. Tech. Eng. News., vol. 8, no. 1, Feb., 1927, pp. 8-9 and 42, 5 figs. Metal is alloy of cobalt, chromium, and tungsten, with any iron present only as impurity; stellite's hardness is inherent, requiring no tempering or heat treating; it is also

highly resistant to action of practically all commercial and industrial acids; it is possible to melt metal by means of oxy-acetylene torch, fuse it to all grades of steel and practically all of cast iron; stellite.

**WELDING APPLICATION.** Stellite: A New Welding Process, A. V. Harris. Am. Welding Soc.—Jl., vol. 6, no. 1, Jan., 1927, pp. 14-17. Stellite is neither welding nor brazing in ordinary sense of words; it calls for blowpipe flame which contains fairly large excess of acetylene to lower flame heat and to exclude as much atmospheric oxygen as possible; stellite can be welded with electric carbon or metallic arc process, but metal resulting from oxyacetylene weld is cleaner, less porous and unalloyed; resistance to heat applications.

### STRUCTURAL STEEL

**BEAMS AND COLUMNS.** New Carnegie Structural Beams. Iron Age, vol. 119, no. 7, Feb. 17, 1927, p. 505, 1 fig. Structural steel beams and column sections in wide variety are covered in new series, descriptions of which have just been issued by Carnegie Steel Co. in 50-page pamphlet; one of chief characteristics of new sections lies in fact that flanges are of uniform thickness from edge to fillet.

**CORROSION.** Is Corrosion of Structural Steel a Serious Matter? Contract Rec., vol. 40, no. 51, Dec. 22, 1926, pp. 1214-1216. Review of comprehensive investigation by F. Skinner, reported to Am. Inst. of Steel Construction, indicates that protection for steel buildings is 94 per cent perfect.

**CORROSION.** Unlimited Potential Durability of Structural Steel, F. W. Skinner. Brooklyn Engrs.' Club—Proc., vol. 25, Jan., 1927, pp. 82-135. Result of researches carried out by author, to show why, how and when structural steel may be corroded and how it is practical to entirely prevent its corrosion.

**RESEARCH.** Research in Structural Steel Design, M. S. Ketchum. Contract Rec., vol. 41, no. 6, Feb. 9, 1927, pp. 129-131 and 139-140. Relatively little comprehensive investigational work in connection with use of structural steel; present status of research in various fields. From paper read before Am. Inst. of Steel Construction.

### THERMIT WELDING

**RAILS.** Experimental Bases for Thermit Welding of Rails on Open Lines (Experimentelle Grundlagen der Thermit-Schienschweißung auf freier Strecke), A. Wöhr. Organ für die Fortschritte des Eisenbahnwesens, vol. 82, nos. 1 and 2, Jan. 15 and 30, 1927, pp. 1-7 and 21-25, 21 figs. partly on supp. plates. Calculation of coefficient of expansion; elongation of rails; costs and economy of welding.

### TUBES

**WROUGHT-IRON, SPECIFICATIONS.** Specifications for Tube Strip and Tubes. Gas World, vol. 86, no. 2221, Feb. 26, 1927, pp. 184-185, 1 fig. Specifications as approved by Council of Institution of Gas Engineers for wrought-iron gas, water and steam tube strip, tubes and fittings.

### TUNGSTEN

**BAR.** Lowering (Sur la barreux P. Wool Bul., vol. 1708-1710 for wire passed t of hyd (3655 and bar to be rounding and sub outside to results

**CRYST.** ence of Property den Ein Furnbes pern), F vol. 7, Tungster few lon small-cry their m ture (th ing), bu in mean of such lamps.

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**TUNGSTEN**

**BARS, HOLLOWING PHENOMENA.** Hollowing of Tungsten Bars by Central Fusion (Sur le phénomène de l'évidement de barreaux de tungstène par fusion centrale), P. Woog. Société Chimique de France—Bul., vol. 39-40, no. 12, Dec. 1926, pp. 1708-1711, 1 fig. In preparation of tungsten for wire drawing, very heavy currents are passed through bars of metal in atmosphere of hydrogen; occasionally melting point (3655 deg. abs.) is accidentally exceeded and bar ruptures, when its interior is found to be hollow, and metal immediately surrounding hollow shows evidence of melting and subsequent recrystallization; cooling of outside layers must be due both to radiation and to conduction and convection by gas; results of test.

**CRYSTALLINE STRUCTURE.** The Influence of Crystal Structure on Form-Retaining Properties of Tungsten Lamp Units (Ueber den Einfluss der Kristallstruktur auf die Formbeständigkeit von Wolfram-Leuchtkörpern), F. Koref. Zeit. für technische Physik, vol. 7, no. 11, 1926, pp. 544-547, 6 figs. Tungsten wires, consisting of a single or a few long crystals, are superior to usual small-crystalline units, not only because of their mechanical behavior in low temperature (they do not become brittle after heating), but because of their constancy of form in incandescent heat; points out importance of such form-retaining wire for incandescent lamps.

**PIT FORMING.** Grain Boundary Phenomena in Tungsten Filaments, E. S. Davenport. Am. Inst. Min. & Met. Engrs.—Trans., no. 1635-E, Feb., 1927, 15 pp., 25 figs. Certain metallic filaments, when subjected to high temperatures, either in vacuum, or in protective or reducing atmospheres, develop small pits or voids in grain boundaries; these pits, for most part, are disconnected and approximately spherical; pit-forming tendency has been most pronounced in samples of tungsten, molybdenum and few other refractory metals made from powders; previous chemical and thermal treatment of metal has pronounced effect on readiness with which intergranular pits develop; control of such variables may give practical means of limiting pitting; size of pits is function of time of burning and probably of temperature.

**WELDING**

**CAST ALUMINUM.** How to Weld Cast Aluminum. Brass World, vol. 23, no. 2, Feb., 1927, pp. 47-48, 4 figs. Difficulties encountered and how to overcome them.

**CAST IRON.** Autogenous and Electric Welding of Cast Iron. Metallurgist (Supplement to Engineer), Feb. 25, 1927, pp. 19-21, 2 figs. Review of work by P. Schimpke, published in Stahl u. Eisen, Aug. 26, 1926, giving comprehensive study of modern welding practice with particular reference to cast iron; includes sketch of moderately large autogenous plant, and points out that low-pressure acetylene generators formerly employed are largely superseded by medium or high-pressure generators working at pressures ranging from 3 ft. of water to one atmosphere; discusses factors which have to be considered in electric arc welding.

**FLOW OF METAL.** The Flow of Welding Metal, J. B. Green. Welding Engr., vol. 12, no. 2, Feb., 1927, pp. 25-29, 13 figs. Slow-motion movies, taken with infra-red light on special film, are used to show welding qualities of filler rods.

**POWER PLANTS.** Special Uses of Welding in Westport Station. Power, vol. 65, no. 7, Feb. 15, 1927, pp. 249-250, 5 figs. Welding used for variety of purposes, including pipe line fabrication, replacement of studs in boiler handhole plates, repair of broken stoker cranks and removal of generator rotor end rings.

**RODS.** Effect of Surface Materials on Steel Welding Rods. Boiler Maker, vol. 27, no. 2, Feb., 1927, pp. 51-52, 1 fig. Discussion of effect on working qualities of metallic and non-metallic surface materials; gas filler rods.

**STRUCTURAL STEEL.** Welding Progress in Structural Field, W. Spraragen. Can. Machy., vol. 36, no. 27, Dec. 30, 1926, pp. 177-207, 48 figs. Demonstrates reliability and economy of welded joints, and their application to structural steel field.

**WELDS**

**FATIGUE RESISTANCE.** Suggested Program for an Investigation of the Fatigue Resistance of Welds, American Bureau of Welding, H. L. Whittemore. Am. Welding Soc.—Jl., vol. 6, no. 1, Jan., 1927, pp. 21-24. Failures of steel members of machine which was subjected to many wide variations in stress were not satisfactorily explained until investigation of fatigue properties of material showed that maximum stresses were higher than endurance limit; tentative results recently obtained indicate that endurance limit of welds may be about one-half that of base metal; rotary beam machine of Farmer type is well adapted for testing welds; other suitable machines.

**PHYSICAL PROPERTIES.** Physical Properties of Welds Produced by Oxyacetylene or Electric Methods (Ueber die physikalischen Eigenschaften der mittels Acetylen-Sauerstoff oder auf elektrischem Wege geschweissten Erzeugnisse), W. Hoffmann. Autogene Metallbearbeitung, vol. 20, nos. 2 and 3, Jan. 15 and Feb. 1, 1927, pp. 18-27 and 33-37, 29 figs. Physical properties of welded joints, including strength and notched hardness; quality tests; welding rods and electrodes; welding of castings; chemical phenomena in connection with welding; cold and hot welding; hot welding of cast iron. Bibliography.

**WIRE**

**ANNEALING.** Use of Salt Baths in Annealing of Wire. Fuels & Furnaces, vol. 5, no. 2, Feb., 1927, p. 241, 1 fig. Coils of wire weighing 200 to 300 lb. are immersed in salt bath until annealing temperature is reached, then removed, permitted to cool and rinsed.

**STEEL.** Testing of Steel Wire (Nouvelle méthode d'essai des fils d'acier). Génie Civil, vol. 90, no. 2, Jan. 8, 1927, pp. 41-45, 24 figs. Author condemns ordinary tensile test suggesting that load under which wire breaks, divided by original cross-sectional area of wire, giving figure usually termed ultimate stress, has little value; in his opinion, elastic limit should be deter-



mined, and suggests simple way of doing this; he attaches great importance to detection of central segregation in steel wire and suggests that wire makers prefer segregated steel; recommends determination of limit of proportionality in tension, shock test, and macro-etching test applied to cross section of wire. See also review in Metallurgist (Supp. to Engineer), Feb. 25, 1927, pp. 31-32.

#### ZINC

**SINGLE CRYSTALS, PREPARATION.** Preparation of Metallic Single Crystals and Twinning in Zinc and Zinc Single Crystals, O. E. Romig. Am. Inst. Min. & Met. Engrs. —Trans., no. 1631-E, Feb., 1927, 23 pp.,

26 figs. Bridgeman method of preparing single crystals; Czochralski and Gomperz method; single crystals of metals may be prepared (1) by annealing at rather high temperature to promote grain growth; often slight straining before annealing facilitates this growth; (2) by drawing polycrystalline wires at determined speed through sharp temperature gradient, as in preparation of single crystals of tungsten; (3) by starting solidification of metal in capillary tube and gradually increasing size of tube; (4) by drawing molten metal through sharp temperature gradient which freezes metal quickly; single crystals possess unusual and different properties from ordinary polycrystalline samples; their physical properties vary with difference in orientation.

### STRESSES IN QUENCHED AND TEMPERED STEEL

By S. L. HOYT

(Continued from Page 530)

volume change sets up a stress, or alters the internal stress, at that section, which is not balanced by the stress at other sections and this results in a strain. This strain is superposed upon the volume change to produce the irregularities recorded. By using Scott's data I have found that this picture is consistent with the experimental facts, that is, that I can predict the way in which the length and diameter will vary during quenching and tempering.

The second last paragraph of Professor Jasper is not clear to me, but if he will indicate the particulars which he has in mind, I shall be glad to attempt a fuller explanation. Unfortunately even a complete application of the present mathematical theory of elasticity will not be sufficient for our present purposes, for, as Love himself points out, it does not hold rigidly for elastic after-working, etc., with which we are directly concerned.

To Mr. Lessells I would say that I made no special effort to keep the bar straight. It warped some, but the measurements were made along four lengths at 90 degrees. The variations in these measurements were not large in comparison to the length changes involved. The attempt to correlate the Brinell hardness and internal stresses is very interesting and I regret not having something to add. The evidence is that the bar with internal stresses gives a point which does not fall on the B. H.—tensile strength line. Why not try to associate this with the effect of the stresses on the tensile strength?

Mr. Klopsch calls attention to the fact that the outside is converted into martensite before the inside passes through Ar'. I do not infer that he implies that this materially alters the situation, for it is the shrinkage of the inside due to the temperature change which exerts the effect. The expansion at Ar' would release the stress comparatively little as the expansion is not large compared to the temperature shrinkage.

Mr. de Forest calls attention to some important effects, and I realize that the stresses studied are not the only ones present. But Scott and Storey have studied these effects, and I would prefer to refer to their work rather than attempt a reply.

The point raised by Mr. Davis has not been properly answered yet. It would be interesting to have the case thoroughly studied and reported upon.

## News of the Society

### A NEW CHAPTER

A NEW chapter of the American Society for Steel Treating located at Dayton has established a record in the history of Society activities, by beginning its existence with an initial membership of 107 members. Of this number 14 are of the sustaining class; 2 are associates, and the remaining, of the member class.

It was when H. H. Skinner, formerly member of the New York chapter in Poughkeepsie, became associated with the Dayton Power and Light Company as gas sales manager that the preliminary work accomplished by W. H. White, chairman of the membership committee of the Cleveland chapter and his Dayton committee, consisting of Otto Klopsch, Charles Merrell, and George J. Oswald, began to bear fruit.

Following a request to the National Office, H. K. Briggs, assistant secretary, spent some time in Dayton assisting Messrs. Skinner, Sisco, Reiter, Klopsch, Oswald and other Dayton members in getting the preliminary work accomplished. Dayton presented a petition, signed by 85 prospective members, to the National Board for a chapter charter, which was granted.

The first meeting was held in the splendid auditorium of the Dayton Engineering Society's building, on Thursday evening, March 24. Mr. H. H. Skinner, chairman of the Organization Committee, called the 250 guests and members to order and gave a brief outline of activities during the organization period and presented the report of a Nominating Committee, which report was unanimously approved.

George J. Oswald, Research Dept. National Cash Register Co., the newly-elected chairman, took charge of the meeting and introduced assistant secretary Briggs and secretary Eisenman, who presented a charter to the chapter. Seventeen additional applications were turned in at this time from those who wished to be considered charter members of the Dayton organization.

After the preliminaries of the organization, Frank R. Palmer of the Carpenter Steel Company presented his talk on "Giving Steel Tools a Chance," which was given in his characteristically interesting style and was most cordially received by all in attendance.

Announcement of the next two meetings was made, the one on Monday, April 18, to be addressed by W. S. Bidle, President of the W. S. Bidle Co., Cleveland; and past president of the American Society for Steel Treating. The May meeting will be addressed by Earl C. Smith, assistant general superintendent of the Central Alloy Steel Corporation, Canton.

The newly-elected officers of Dayton chapter are as follows:

CHAIRMAN—George Oswald, National Cash Register Company

VICE-CHAIRMAN—O. Z. Klopsch, Delco-Light Company

SEC'Y & TREAS.—F. M. Reiter, Dayton Power and Light Company

## EXECUTIVE COMMITTEE—

E. C. Carter, National Cash Register Company  
H. H. Skinner, Dayton Power and Light Company  
R. G. Rodgers, Delco-Remy Corporation  
H. M. Williams, Delco-Light Company  
T. J. Mullen, City Machine and Tool Company  
C. E. Merrill, Gustave Wiedeke Company  
F. T. Sisco, McCook Field

The Dayton chapter at their own request start with their chapter listed in Group II. This is indicative of the splendid spirit and enthusiasm permeating the members of this new division, and also that the successful inauguration is but a slight indication of greater successes that will continue in the future.

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GENERAL INDEX

THERE is in process of preparation a general index to all of the material so far published by the American Society for Steel Treating. The publications being indexed include the 10 volumes of the *TRANSACTIONS*, 2 volumes of the *Journal* of the American Steel Treater's Society and the 2 volumes of the *Proceedings* of the Steel Treating Research Society, a total of approximately 12,000 pages.

Mr. Frank T. Sisco, McCook Field, Dayton, the well known author and metallurgist, has been engaged for about one year in the preparation of this index, and it undoubtedly will be as comprehensive, complete and satisfactory as it is possible for an index to be. It will contain approximately 300 pages 6 x 9, and will be in the same binding as at present used for the bound volumes of the *TRANSACTIONS*.

The cost of preparation and publication of 1000 copies of this index will be approximately \$4000.00, or at the rate of \$4.00 per index. However, inasmuch as the index is published as a service to the members of the Society, the Board of Directors has authorized the sale of the index at the rate of \$1.00 per copy providing orders for the index are received in advance of the date of publication.

Proper order forms for the index will be mailed to the members with the next issue of data sheets.

## News of the Chapters

### BOSTON CHAPTER

THE March meeting of the Boston Chapter was held at the Massachusetts Institute of Technology on Friday, March 4, 1927. Dinner was served to ninety members and guests in Walker Memorial and about one hundred and fifty attended the evening meeting. Among the guests were C. M. Freitag of Providence Chapter and our ever-welcome guest A. H. d'Arcambal of Hartford.

Following a short business meeting at which a summary of the membership gains and losses for the last two years were given, Dr. G. B. Waterhouse gave one of his interesting talks on current events in heat treating. He selected as his subject a recent paper on the growth of cast iron after repeated heatings, which was very interesting due to its relation to carburizing pots and furnace parts. Pieces 6 inches long and 0.6 inches in diameter were prepared, then heated to 1650 degrees Fahr. and cooled fifty times. Five different irons were taken, the silicon being different in each, the balance of the elements being similar in each sample. The following results were obtained:—

- 1—Iron with 0.68 per cent Silicon grew 6 per cent
- 2—Iron with 1.06 per cent Silicon grew 17 per cent
- 3—Iron with 1.50 per cent Silicon grew 24 per cent
- 4—Iron with 1.99 per cent Silicon grew 29 per cent
- 5—Iron with 2.50 per cent Silicon grew 40 per cent

It is apparent from the above that annealing pots and furnace parts should not be made of high silicon cast iron. It was also noted in the paper that the growth was much retarded by heating in hydrogen and accelerated by heating in carbon dioxide, showing that the growth was in proportion to the amounts of oxygen present.

The principal speaker of the evening was F. B. Lounsberry, vice-president of the Atlas Steel Corporation, Dunkirk, New York, who selected as his subject "Tool Steel Failures, Their Causes and Cures".

The talk was divided into three parts:—

- I —Why did it fail?
- II —Whose fault was it?
- III—Suggested remedies.

Mr. Lounsberry also discussed the various things which assist the tool maker in the selection of the proper steel for a given job. As a means to obtain the greatest possible service from the manufactured tools, he suggested more co-operation between the tool steel manufacturer and the customer; and in the customer's plant itself, more co-operation between the purchasing department, the tool room and the hardener.



The talk was fully illustrated and in a language which could be readily understood. In addition to a large collection of micrograph slides, Mr. Lounsberry showed many illustrations of fractures of defective tools and dies. One of the slides shown was a milling cutter with several teeth broken out. Our visitor from Connecticut volunteered the information that this was known as a "pyorrhea cutter". Considerable discussion accompanied and followed Mr. Lounsberry's very interesting talk.

A letter has been received from the chairman, F. H. Cole, whose work keeps him in New York a greater part of the time. He regrets that he has had to miss a couple of meetings, but hopes to be with us occasionally.

The annual meeting will be held this year at Beverly, Mass., on May 6. The United Shoe Machinery Corporation has invited us to spend the day in its plant, and after dinner at its Country Club. J. C. Keilman, of the New Departure Manufacturing Company, will address the chapter on "Forging By the Upset Process". Members of other chapters are cordially invited on condition that they notify the secretary of the Boston Chapter of their intention to be present. Bring along your golf sticks and a good right arm for horseshoes. We hope to have the national officers with us on this occasion.

H. E. Handy.

#### CHICAGO CHAPTER

Before a record breaking crowd taxing the capacity of the City Club, F. B. Lounsberry, vice-president Atlas Steel Corporation, Dunkirk, N. Y., spoke before our chapter Thursday evening, March 10. Mr. Lounsberry's subject was "Tool Steel Failures—Their Causes and Cures." The paper was illustrated with lantern slides and brought out clearly the subject from the tool steel manufacturers' standpoint as well as from the standpoint of the tool steel users.

The meeting opened by the introduction of our new president, J. Fletcher Harper, who spoke briefly upon the "Past, Present and Future of the American Society for Steel Treating."

This was followed by a very humorous talk by our genial national secretary, "Bill" Eisenman, pertaining to his trip abroad.

#### CLEVELAND CHAPTER

F. R. Palmer of the Carpenter Steel Company, Reading, Pa., was the speaker at the meeting of the Cleveland Chapter held on Friday evening, February 18, 1927. The subject of his address, "Giving Steel Tools a Chance", is one which is of vital interest to men engaged in the steel tool industry. His talk covered the field of tool design, cause of tool failure, preferred heat treatment, etc. Many new and interesting points were brought out concerning the practices just mentioned. Mr. Palmer presented his subject in a delightful and interesting manner, making it understandable and of practical value to the shopman as well as to the executive.

On Monday evening, February 28, 1927, the Cleveland Chapter held a special meeting at Physics Building, Case School of Applied Science. The speaker was Professor C. H. Desch of Sheffield University, Sheffield, England, and he addressed the members on "Modern Views on the Deformation of Metals".



Professor Desch is a widely known English metallurgist and the author of several books on metallography, alloys, the phase rule and others. The speaker stated his subject in an unusually interesting form and was very well qualified to present this type of subject.

"Quenching and the Quenching Efficiency of Various Coolants" was the subject of the address given by O. Z. Klopsch of the Delco Light Company, Dayton, O., on Friday evening, March 15, 1927, to members and guests of the Cleveland Chapter.

Mr. Klopsch was exceptionally well qualified to discuss this subject and brought out many interesting facts concerning various tests which H. J. French and he had made at the Bureau of Standards, Washington, D. C. His talk was accompanied by lantern slide charts. The numerous questions put to him by members of his audience were ably answered by the speaker.

Mr. Eisenman seemed to think an investigation should be made on the old superstitious methods of quenching, and it was voted that he may conduct these at any time to his own satisfaction.

About 125 members were present at each meeting. *A. E. Buelow.*

#### DETROIT CHAPTER

The February meeting of the Detroit Chapter struck a new note in subjects. These were:

"Pickling Methods" by Professor E. M. Baker of the University of Michigan.

"Sand and Shot Blasting" by L. Birckelbaw of the Hudson Motor Company.

"Eletero Cleaning" by W. M. Graves of Packard Motor Company.

The evening was well started and exceedingly amusing due to the "Coffee Talk" by C. C. Bradner. He can be well described by quoting from the announcement of the meeting:

"Mr. C. C. Bradner occupies one of the inclusive columns of the Free Press. He is one of the reasons we pay three cents for the paper. Mr. Bradner will talk on something else, some of the time, and concerning assorted nuts the rest of the time. He did and we sure laughed off our troubles.

That these unusual subjects for steel treating society meetings were well planned is attested by the lively discussion and interested questions that they drew.

The attendance at dinner, as well as at the meeting, was as usual about seventy-five and one hundred twenty-five for the meeting following.

Professor Baker's talk on "Pickling Methods" was a brief description of the steps in the operation of pickling, both of sheet stock and heavier forging. He explained the differences in performance of various solutions, particularly pointing out the effect of acidity, concentration and conductivity of the pickle as well as some probable reasons for effectiveness of some inhibitors. These interesting additions and discussion showed their general use.

Mr. Birckelbaw's talk showed an unusual analysis of the common operation, sand blasting. He described the mechanism of various materials for these operations and gave some illustrations of how to analyze cost figures to indicate comparative values.

Mr. Graves' talk described the success of a method of "cathodic" electro pickling. The quality of the cleaned work was well demonstrated by a sample gear brought in to the meeting.

J. L. McCloud.

Courses in metallurgy are being given in the evening schools of the College of the City of Detroit and the Cass Technical High School.

The evening courses in metallurgy given at the College of the City of Detroit consist of a study of metallography and fuels and their utilization.

Metallurgy I and II are being given at Cass Technical High School. Metallurgy I consists of the study of the making of pig iron, the different processes of making steel, its working and heat treatment. The use of the microscope is taught in this course. Metallurgy II is a continuation of Metallurgy I. Solving problems that occur in the shop and the practical use of the microscope are given attention. Fee for the courses is \$4.00 each.

#### GOLDEN GATE CHAPTER

An elementary, practical course of twenty illustrated lectures on the subject of "Steel and Its Heat Treatment" was begun on March 8, 1927, by Golden Gate Chapter in collaboration with the Oakland Board of Education.

The course will cover a brief introduction to the history of metallurgy (ancient and modern), the principles of chemistry, and the manufacture of steel from the raw material to the finished heat treated product. Consideration will also be given to the methods of testing and inspection.

The lectures will be fully illustrated with the reflectoscope. H. S. Taylor and C. S. Moody, metallurgical engineers of the Caterpillar Tractor Company, are the instructors. There is no tuition charge, but a nominal charge will be made to cover the cost of printing the notes. These will be illustrated with about 135 cuts and will be worth keeping. The course is open to anyone. Additional information may be obtained from the secretary of Golden Gate Chapter.

#### HARTFORD CHAPTER

The February meeting of the Hartford Chapter was held on the 8th in the Assembly Hall of the Hartford Electric Light Company. The speaker was Professor H. F. Moore of the University of Illinois. He delivered a paper on "What Happens When Steels Fail by Fatigue."

Professor Moore indicated that fatigue failures are progressive rather than instantaneous over the entire surface. Under repeated stress, this progress can sometimes be followed. A moving picture film illustrated this point.

In explanation, he indicated that apparently there is a tearing apart of the metal particles even though the repeated stresses might be well below the elastic limit, with which fatigue limit has no relation. This is due to the fact that the metals are not homogeneous and contain local areas of weakness. These will give first and the spreading follows. The local weakness might be put into the piece in question as for example through sharp corners.

For the so-called engineering steels, structural steels up to about 450 Brinell, it has been found that loads up to 50 per cent of the ultimate strength

or about 250 times the Brinell hardness will usually resist fatigue for an indefinite period.

Professor Moore's excellent delivery together with the subject found instant favor with the large gathering. A lively discussion followed the presentation of the paper.

*H. J. Fischbeck.*

#### LOS ANGELES CHAPTER

The last regular meeting of the Los Angeles Chapter of the American Society for Steel Treating was held February 12, 1927, at 2:30 p. m. in the plant of the Douglas Aircraft Company located at Santa Monica, Calif.

Through the courtesy of the Douglas Company officials, the members of the Society and their friends numbering about one hundred in all had the rare privilege of an inspection trip through this, one of the most modern, complete, and up-to-date aircraft manufacturing plants in the United States.

Mr. Kindleberger, chief engineer of the Douglas Company, escorted the visitors through the various departments explaining in detail the various phases of construction of the planes as we passed through the departments.

The plant consists of five large ship buildings where all of the Douglas planes are built. All parts of the ships, with the exception of the motors and instruments, are made and assembled here in this plant.

Mr. Kindleberger gave a very interesting talk on the welding of the special metals and steel tubing used in the construction of the fuselage of the planes. Stress was also placed upon the accuracy of the jigs used to take care of expansion and contraction of parts used in the manufacture of the planes.

The members were very much interested in the great amount of handwork in making up all of the laminated wooden parts going into the planes and the accuracy with which every part is shaped. It was interesting to see the large wooden members built up from so many thousands of small pieces each accurately shaped.

The wing shops where all wings were made proved very interesting as we were able to see how the special airplane cloth is shrunk, weather-proofed, painted and assembled.

Finally, we were shown the assembly shop where the complete planes are assembled; where the motors and all instruments are mounted, and, if government planes, where the guns are installed. Two, nine-men capacity planes were ready for shipment at the time of our visit.

Such keen interest was shown throughout the various departments that Mr. Kindleberger suggested we make an annual affair of a plant inspection through this plant.

After a vote of thanks to Mr. Kindleberger the meeting adjourned.

*E. C. Black.*

#### MILWAUKEE CHAPTER

The Milwaukee Chapter held its February meeting on the 14th at the Republican Hotel.

A banquet at 7:00 p. m. with an attendance of fifty was followed by a lecture on high speed steel, its structure, composition, and treatment. Two reels of pictures illustrating operations in a tool steel mill, lantern slides and

fractured specimens were an interesting part of a very instructive lecture.

J. P. Gill, metallurgist of the Vanadium Alloy Steel Company, Latrobe, Pa., was the speaker of the evening to whom we wish to express our thanks, also to the Vanadium Alloy Steel Company. *M. E. Greenhow.*

### NEW HAVEN CHAPTER

When the writer walked into the dining room of the Hotel Bishop and saw so many smiling faces, he was at once convinced that the dinner meetings have come to stay, and, from now on, will occupy a permanent place with the New Haven Chapter.

After having the inner man satisfied, we journeyed to the Hammond Laboratory of Yale University for the regular meeting.

During the business meeting, C. J. Sauer, our honorable chairman, told the members and guests that this was the last meeting of the year to be held in New Haven, and on behalf of the chapter thanked the faculty of Yale for the use of Hammond Laboratory for our winter meetings. It has indeed been a pleasure to meet under such pleasant surroundings.

It seems about this time of the year the New Haven Chapter gets the wandering fever and wants to do something different, so in April we are going to wander to Bridgeport for our regular meeting, and in May we are going to Waterbury, winding up our year in June with our Annual Frolic at Savin Rock Park. By the way, have you ever been to our June meeting? If not, you have missed something. If you do think so, plan to be on hand this year because the chairman of the Frolic has promised that it will surpass all others. Nuf sed. Ask Walter.

After the short business meeting, the chairman announced that our speaker, W. H. Dunbar, assistant general sales manager of the Norton Company, had been unexpectedly called away on business, but that E. C. Hughes, sales engineer of the same company, had been sent in his place. Judging from the very interesting talk delivered by the speaker, I would be safe in saying that the substitution was very good.

A brand new four reel motion picture, "The Age of Speed," made by the Norton Company, showing the part grinding plays in our present day existence was shown. This film was historic, educational, and entertaining and contained some very interesting technical information.

Following the meeting, the chairman threw the meeting open for discussion, and from the number of questions asked, it seemed as if everybody has grinding troubles. We are indeed glad that the subject brought forth a good discussion because the chairman of the program committee worked hard this year to provide a program which would prove interesting. Good work, Paul, for your efforts and judging from the attendance this year, the members seem well pleased.

The attendance prize donated by W. Paul Eddy, Jr., of the Geometric Tool Company was won by J. M. Dillon of the same company.

We had a very good attendance of sixty-five, but many of the old familiar faces were among those missing, especially the steel salesman, but at the same time, we were glad to see so many new faces. Among those missing were



Dawless, Duley, Richardson, Besom and Bellis. Well, don't let it happen again.

The Porter family, father and two sons, were with us, we are glad to say; in fact, Bridgeport was very well represented. R. Wallace & Sons Manufacturing Company, Wallingford, were represented by nine men; Geometric Tool Company, New Haven, had seven men out, and Heppenstall Forge Company, Bridgeport, had six men out. Congratulations to these companies. Keep the good work up. I hope some of the other firms will come to the front before the end of the year.

To miss a meeting of the New Haven Chapter this year is missing an opportunity to hear something worth while. Some of our members have been to every meeting.

*W. G. Aurand.*

#### NORTHWEST CHAPTER

The March meeting of the Northwest Chapter was held Wednesday evening, March 16, at which R. S. Archer gave a lecture on Aluminum and Aluminum Alloys. The meeting was preceded by a dinner at which about twenty members were present.

Mr. Archer first discussed briefly the history and development of aluminum metallurgy and then gave in detail the Hall and Hoop processes which are used at the present time. The low purity aluminum is used chiefly as a deoxidizer in the steel industry while the high purity metal is used in electric cables, cooking utensils, foil for food wrappers, paints, collapsible tubes to replace tin, and for chemical equipment as aluminum is very resistant to certain kinds of corrosion.

The casting alloys of aluminum were next discussed, especially as regards their ease of casting, method of casting, physical properties and uses. The three alloys of chief importance in this field are the copper, zinc, and silicon alloys.

Mr. Archer next considered the wrought aluminum alloys, giving the various compositions used and their properties. He also gave the heat treatment for these alloys and an explanation for the changes in properties in these alloys due to the heat treatment.

After the lecture there was a good discussion on the strength of the aluminum alloys especially as compared with steel.

This chapter was extremely fortunate to have Mr. Archer present this interesting subject because of its great importance in engineering work at the present time.

*L. J. Weber.*

#### PITTSBURGH CHAPTER

The Pittsburgh Chapter held its March meeting on the evening of the 3rd at the U. S. Bureau of Mines. Following the usual executive meeting, a dinner was served in the Bureau of Mines Restaurant at 6:30 p. m., which was attended by about fifty members. At eight o'clock, Chairman W. H. Phillips called the meeting to order in the auditorium and during the short business session, Secretary H. R. Wade distributed the new membership list. This list is in the form of a leather bound booklet and in addition to the names of the Pittsburgh Chapter members, the national officers and the year's



program, it contains a photograph and a short biography of Dr. Frederick Crabtree, the chapter's first chairman.

At the conclusion of the business session, the chairman welcomed H. J. French, senior metallurgist of the U. S. Bureau of Standards at Washington, D. C., as the speaker of the evening. The subject of Mr. French's talk was, "A Comparison of the Alloying Elements Chromium, Nickel, Molybdenum and Vanadium in Structural Steels," and it proved to be a talk replete with detailed information which was exceedingly valuable and interesting. Mr. French introduced his subject by giving the effect upon steel of the above mentioned four alloying elements. He then divided their use into two general groups—the low alloy and the high alloy types.

In the low alloy group were nickel, chromium, chromium-vanadium, chromium-molybdenum and nickel-chromium steels and he compared them in respect to their making and shaping, their machining properties, their heat treatment properties, and their welding, cold-working and carburizing properties. In the high alloy group, he compared steels containing much higher percentages of these four alloying elements in respect to their ability for resisting corrosion and for high temperature service. He then took up alloy steels for resisting wear and gave some interesting information though, as he said, so little co-ordinated information is available that it is hazardous to make any very definite or general statements.

Mr. French concluded his talk with a general summary of the subject. Throughout this most interesting talk, lantern slides were used which showed charted information relative to the various phases of the subject, and also photographs of the actual uses to which the various steels discussed had been put. A broad discussion of this subject was then participated in by a number of the members until the lateness of the hour made adjournment necessary and it was done with a hearty applause in appreciation of Mr. French's splendid and very comprehensive talk.

A more detailed report of this paper has not been made as it will be published in its entirety in the TRANSACTIONS.

*Harry A. Neeb Jr.*

#### ST. LOUIS CHAPTER

The sixty-fourth monthly meeting of the St. Louis Chapter of the American Society for Steel Treating was held Thursday evening, February 17, 1927, with 160 members and guests present. This was one of the largest attendances ever recorded for the St. Louis Chapter. It was indeed gratifying to note the new faces from the old sustaining and the new sustaining members. We want to thank our sister cities, Alton, Ill., and Granite City, for their excellent turnout. Mr. Myers of the Western Cartridge Company attended in full force with his associates, also we noted new faces from the Laclede Steel Company, Alton, Ill., and it was gratifying to see our new sustaining member—namely, the Century Electric Company, appearing in full force after joining our Society.

After the dinner the meeting was called to order by Mr. Swander, chairman, and the secretary announced several new members. After a very short meeting the speaker of the evening, G. A. Richardson, manager of the technical

April

publicity department, Bethlehem Steel Company, was introduced. The title of his lecture was "The Manufacture of Wire and Wire Products at its Maryland Plant, Sparrow's Point, Md., and at the Original Mill at Johnstown."

There were eight reels of moving pictures, which are considered the finest series which has ever been produced by their technical publicity department. The showing was one of the first, and a very recent release. The picture showed some very unusual scenes, showing Bessemer converters, the rolling of the billets and the work of the various departments of their wire plants. In fact, every step in the manufacture of wire and wire products was shown, starting at the continuous heating, furnaces, passing to the continuous rod mills, the new muffle type conveyors, cooling racks, cleaning houses, wire and annealing division, and the manufacture of special parts, such as woven wire fence, wire nails, galvanized wire, barbed wire, etc. Mr. Richardson's lecture carried him all the way through the presentation of the subject, which afforded our membership and guests an evening's instructive entertainment which, in the writer's opinion, was well worth the time of every one present.

There being no further business, Mr. Richardson was given a rising vote of thanks, and the meeting was adjourned until March.

We wish to advise that F. C. Langenberg of the Watertown Arsenal will lecture before the St. Louis Chapter in March, and in April we will have A. H. d'Arcambal of the Pratt & Whitney Company; and for May we have Dr. J. A. Mathews of the Crucible Steel Company—which should be a great inducement for the attendance of the St. Louis Chapter. *C. G. Werscheid.*

#### SOUTHERN TIER GROUP

The February meeting of the Southern Tier Group was held in the cafeteria of the Willys-Morrow Company, Elmira, N. Y., February 23, 1927. Sixty-five members and their guests were present at dinner. Their number was swelled by others unable to attend the dinner but attending the talk afterwards bringing the total attendance for the evening to about ninety persons. Binghampton, N. Y., Endicot, N. Y., Painted Post, N. Y., and Athens, Pa., were represented by groups of from three to twelve men.

C. E. Killinger, vice pres. and general manager of the Willys-Morrow Company, spoke a few words of welcome to the members of the society and their friends. He told how he has watched the growing importance of the metallurgist and the steel treater during the past twenty years until today he is a very important cog in the machinery of manufacturing.

The speaker of the evening was F. R. Palmer of the Carpenter Steel Company, Reading, Pa. He gave his talk "Give the Steel Tools a Chance."

At the beginning of his talk Mr. Palmer called attention to the well-known but seldom thought of fact that when the sum of the strains in a tool exceeds 100 per cent of the strength of the steel, the tool must fail.

Then, starting with the supposition that the steel was of high quality and defining high quality as cleanliness, purity, and uniformity, Mr. Palmer talked of the contributing factors which set up strains in a tool. By means of slides, he illustrated his points in a convincing manner. He took in turn the tool designer, the tool maker, the heat treater and the grinder, pointing out how each one could contribute to the sum of the strains in the tool and repeatedly

referring to the simple truth that when that sum exceeds 100 per cent the tool must break.

Mr. Palmer ended his talk by making a plea to all tool makers and employers to give the heat treater a square deal. Enlarging on this, the speaker said that so often all possible care is taken with a tool or a piece of steel until it reaches the heat treater. Then he is ordered to rush the job when his better judgement tells him that to do so will result in a defective tool. Further than that many companies have wonderful machines and equipment in their shops but the heat treating furnaces are poorly designed and the pyrometer equipment is conspicuous by its inaccuracy, improper application or total absence. To bring home his point, Mr. Palmer asked what the tool maker would say if he were given a blue print marked in decimal dimensions, a piece of steel, a dressmaker's tape measure and told to "Go to It."

### TORONTO CHAPTER

Probably unique in American Society for Steel Treating history was the action of Toronto Chapter in staging a vaudeville revue and euchre, which will now be an annual event, in the Metropolitan Assembly Rooms, 245 College Street, Toronto, December 23, in order that the lives of eighty-seven children of destitute workmen formerly employed in various phases of the metal working industry, might be brightened, Christmas Day.

Suggested at an executive meeting of the chapter by T. H. Young, Jessop Steel Company; at once endorsed by W. J. Blair, Canada Cycle and Motor Company; J. W. McBean, Central Technical School; J. A. Dow, Canadian Acme Screw & Gear Ltd.; Cliff Cornwall, Canada Illinois Tool Company; A. Lowry, Massey-Harris Company; A. G. Davis, Consumers' Gas Company; and Campbell Bradshaw, managing editor, MacLean Technical Newspapers, one of the best "shows" ever held in the Queen City of Ontario was witnessed by a capacity house of steel treaters and engineers, equipment manufacturers and their representatives.

Despite the fact that the benefit was suggested for the first time on December 9, every seat was sold forty-eight hours before "the rise of the curtain".

Granted exemption from the amusement tax by Dr. J. D. Monteith, provincial treasurer of Ontario, the proceeds were devoted to croquet sets, Santa Claus spinners, pork and beans, biscuits, bananas, picture books, onions, coal, coke, charcoal, bread, roast beef, oranges, apples, chocolate bars, potatoes, flour, baking powder, mittens, stockings, undervests, blankets, sheets, tea, candy, mouth organs, butter, sugar, carrots, cabbage, turnips, dolls, and balloons.

The nature of the entertainment can be determined in some measure from the fact that acts from the B. F. Keith time, Shea's Hippodrome; Loew's Theatre, Uptown Theatre, and Leo Feist Inc., the music publishers, donated their services through the courtesy of the various theatre managers.

The publicity was handled by the editorial staffs of Canadian Machinery and Power House, two of the MacLean technical newspaper, and distribution was undertaken Christmas Eve. Preparation of the baskets began at 6 p. m., and distribution was continued without cessation until 4 a. m. Christmas

morning. Work was resumed again at 9 a. m. and completed at 11:30 a. m. Steel treaters to whom especial credit is due for this phase of the undertaking were J. W. McBean, Cliff Cornwall and D. G. MacInnes.

A month later, on January 21, Charles McKnight, Jr., International Nickel Company, repeated his talk on "Some of the Newer Developments on Nickel Steel", which was so enthusiastically received by the Montreal Chapter a week earlier. And again Mr. McKnight emphasized the leadership of Canadian railroads in the application of nickel steel, a product which has its origin in the Dominion, to steam boilers, as a result of the limitations now evident in the weight and size of locomotives.

Mr. McKnight said in part:

"While the heat treatment of alloy steels is a familiar subject to the average heat treater, there are recent developments in the use of nickel steels that are not generally known.

Today there is no field for the use of nickel, except, perhaps, the railroads' growing use of nickel steel, that shows such promise.

To sum up briefly, nickel in cast iron tends to refine the grain; to increase the machinable hardness; to increase the resistance to wear; to increase the strength; to reduce the chill and to eliminate porosity. This list sounds like the collection of ills cured by some patent medicine, especially as some of them seem contradictory. This can be explained by saying that when nickel is to be added to cast iron, the other elements must also be controlled in order to gain the desired end.

The increase in the wear of castings when nickel is used is due, first to this increased hardness, and, secondly to the finer structure and freedom from carbide particles. Under wear conditions, these hard carbide particles act as a lapping compound and increase the wear considerably.

These valuable characteristics are exhibited by the addition of nickel alone in cast iron, but other alloys can be used to supplement the nickel, the most common one being chromium. The conjunction of the two is mutually advantageous. Chromium is a powerful hardener of iron but is also very active in producing chills and carbide spots. This tendency is counteracted by nickel. The best ratio of nickel to chromium is two or three to one.

Cast iron with nickel alone or in combination with other elements, is used today for a wide variety of purposes, such as automobile cylinders and pistons, differential spiders, Diesel engine cylinders, hydraulic press castings, valve and pipe fitting castings, electrical resistance grids, pipe balls, steam cylinder bushings and piston rings and rolls for steel mill service.

*Campbell Bradshaw.*

#### TRI-CITY CHAPTER

The meeting was held on Thursday evening, February 17, at the LeClaire Hotel, Moline, Ill. J. P. Gill, chief metallurgist of the Vanadium-Alloy Steel Company, spoke before the Tri-City Chapter of the American Society for Steel Treating on "Carbon Tool Steel." Mr. Gill's talk was divided into three parts—the composition—method of manufacture and heat treatment.

Under composition was discussed the effect of the different elements ordinarily found in carbon tool steels. Comparison of crucible, electric and



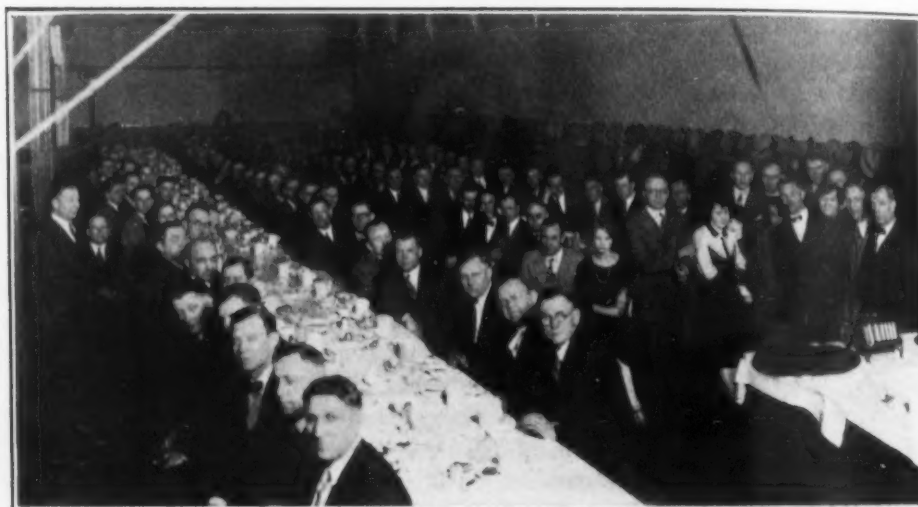
open-hearth steels was explained and the effect of different mill operations on subsequent structure discussed. Effect of time and temperature hardening, with length of time at tempering temperature was explained in detail. The talk was closed with some remarks regarding normal and abnormal steels and in which it was stated that abnormal steels could be hardened without soft spots if proper precautions were taken.

Mr. Gill was accompanied by W. R. Mau, Chicago district manager of the Vanadium-Alloys Steel Company.

#### TRI-CITY CHAPTER

The meeting was held on Tuesday evening, March 1, at the International Harvester Tri-City Works, Rock Island, Ill.

Equipment and operation of the huge tri-city tractor plant of the International Harvester company on Fifth avenue, Rock Island, near the border-



line between Moline and Rock Island were inspected by 150 members of the Moline chapter of steel treaters with Professors Fielding, Fleming and O'Brien of the University of Iowa and twelve engineering students. The technical men were the guests of the International Harvester company at a dinner in the plant cafeteria, following which they were taken through the factory. Arrangements for the dinner and the trip were made by E. H. Sohner, general manager, and J. W. Phillips, assistant general manager. Entertainment at the dinner was furnished by the company orchestra and quartet. The inspection tour through the great factory, described as one of the most modern tractor plants in the world, under the care of guides for each small group of guests was greatly enjoyed by the entire party.

The trip first took the visitors through the new tool room, then down the assembly chain where the tractors are put together as they travel through the shop. After assembly, the tractor is sprayed, washed and painted. The drying is accomplished as the completed machine passes through an oven nearly 300 feet in length. The guests then saw the testing department where the



newly assembled tractors are given their final adjustments and tuning before shipment. The party visited the wheel shop. Here steel wheels for the tractors are turned out at the rate of 200 to 250 a day. An enormous punch press and groups of bulldozer machines, made by the Williams, White & Company of Moline, held the interest of the visitors for some time. A maze of automatic machines for turning out parts of the differential and transmission was viewed by the guests. Automatic gear cutting machines seemed to hold the most interest.

Three gigantic furnaces, each thirty feet in length, marvels of engineering perfection, greeted the party as it entered the heat treatment section of the factory. Electricity and gas are the fuels used for the furnaces. These, the guides explained, give the uniformity of heat and automatic control necessary to treat the most exact parts going into the tractors. There are no men here charging the furnaces and at the same time watching and controlling the heat as is the case in the majority of factories. The temperature of the furnace and the speed with which it is charged and emptied is controlled by the department head in his office overlooking the entire shop. Charts are kept in the office showing and recording continuously the temperature in various portions of the furnace.

One machine in this department claimed more than its share of the attention of the visitors. It was the Ajax upsetting machine, one of the three of its kind and size in use in the world. Huge bars of white hot steel  $3\frac{1}{2}$  inches in diameter were fed into this machine and in less than a half minute came out flattened to a diameter of 14 inches. The press weighs 158 tons and cost approximately \$55,000. Forgings upset in this machine are held to be superior to hammer forgings in that they can be made to closer limits and the operation of trimming flash is eliminated. The visiting party was the first that has made an inspection through the International Harvester factory since its establishment here in the fall.

*Geo. A. Uhlmeier.*

### WORCESTER CHAPTER

The Worcester Chapter held its regular March meeting Thursday, the 10th, at Baratti's Restaurant, Worcester, Mass. Seventy-five members and guests were present at the lecture which was preceded by a supper at 6:30 o'clock p. m.

A few announcements were made by E. D. Clark, our chairman, and Mr. Bigelow of our membership committee, after which Mr. Clark introduced the speaker for the evening, Jordan Korp of the Leeds and Northrup Company. Mr. Korp gave a practical talk on "Tools, Design, Heat Treatment and Its Relationship to 'Production'". A brief outline of this talk follows:

Mr. Korp started his talk by relating the marked difference as to the information given the designer and tool maker as compared to information given the hardener. As a general rule the designer and tool maker have a wealth of information to work with in order to make a perfect tool. In fact, if any dimension is missing it is quickly furnished. All this tends to give the tool maker ample opportunity to make a perfect tool. This perfect tool is then sent to the heat treatment room with information of this char-

acter: "Here is a tool and we want it *right* away!" The hardener who has as many variables as the tool maker to deal with, treats the tool with a meagre amount of information.

The speed that is required because tools are wanted—thus not allowing them enough time in the heat treatment room—is usually disastrous. Speed cannot be successfully practiced in the heat treatment room. The hardener (who is more afraid of his job than spoiling a tool) usually consents to force the hardening of a tool too quickly. The management is the chief offender. Mr. Korp had an opportunity to see a large tool heat treated in fifteen minutes when it should have taken three hours. It was subsequently ruined.

The hardening of steel successfully requires that strict attention be paid to details. The rate of heating of a piece of steel in hardening is not given enough attention. Steel should not be heated faster than it will absorb heat. Quick heating will cause strain, because of the uneven condition of expansion, that often results in failure. The number of degrees that the steel is heated above the critical temperature is important. This will vary with the nature of the work and it should be borne in mind that the rate of heating will vary the critical temperature. Cutting ability and depth of hardness will all be affected by the temperature to which a tool is heated.

The rate of cooling and volume of quench are very important. The faster the rate of cooling the more the strain in a tool. Quenching a large tool in a tomato can is not good practice and the amount of liquid should be regulated with regard to the size of the work to be quenched. The temperature of quenching medium is also to be watched carefully as sometimes a few degrees variation results in a marked difference of properties of quenched steel. In the use of a brine solution the density should be controlled for best results. A saturated brine solution will give soft spots on tools quenched in it. If the use of a jet or spray of liquid is practised the pressure is of importance.

Drawing or tempering of steel should have quite a lot of attention. The effect of time and temperature are of marked importance. Temperature will decrease hardness and time will increase toughness. The drawing of steel by color is obsolete and if used results in very non-uniform condition to the tool.

The hardener gets a hundred per cent of the blame when a tool fails and failure occurs when a tool is not everlasting. It has been proven that the hardener is not always to blame, for the designer and tool maker come in for their share. We must control all conditions if we expect to get the best results and the heat treater should consider all conditions very seriously.

In concluding Mr. Korp had a number of questions to answer, which he did in an interesting manner.

C. G. Johnson.

April

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